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Metals and Alloys

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Making Gears ?.. then

Above — One of the laboratory furnaces used for developing the proper grade and heat treatment for gear steels.

NOTHING IS LEFT TO CHANCE

Precise uniformity of control is exercised in every step in every heat of U·S·S Carilloy Steel produced. It includes uniformity of selection of scrap for furnace charge, uniformity of furnace practice, uniformity of slag control, uniformity of tapping, pouring and rolling practice, and uniformly close metallurgical control, supervision and inspection throughout the above steps until the steel is delivered to your shop.

Below — Ascertaining steel hardenability by macroscopically examining fractured samples and etched sections.

UNITED

editorial

WILLARD GIBBS—and Others

From one of the staff of the publishers of METALS AND ALLOYS comes the following inter-organization memo: "I should have thought that METALS AND ALLOYS would have commented editorially in February anent February 11th being the centenary of the birth of Willard Gibbs: it is generally agreed that modern scientific metallurgy stems from Gibbs' classical work on thermodynamics resulting in the phase rule." The point is well taken. If we'd had to rack our brain for a topic for a February editorial, this would have been a logical subject. As there was already an extra long editorial for February, no racking process was due.

On the whole, we are not personally impressed by anniversaries—e.g. the celebration of the President's birthday for the worthy purpose of getting funds to fight infantile paralysis seems to us to degrade and handicap the project, especially as the papers say that in Columbus out of something like $6500 paid in, something like $2100 was netted. Centenaries are only slightly more intriguing than anniversaries. Still, centenary or no centenary, and belated or not, a salute to Gibbs from metallurgists is in order.

However, we can't forget that Gibbs was satisfied to put his epoch-making mathematics in an obscure publication and, once it was off his chest, to let it lie without any effort to apply it.

At least equal recognition is due to those who dug Gibbs' work up from obscurity and showed its application to practical metallurgical problems. Without the interpreters, Gibbs' work would not have had the acclaim it deservedly had in the February issues of chemical journals. We salute the interpreters also, and equally!—H.W.G.

As Others See Us

From the observation post of London, the British "Metal Industry" of Dec. 9 editorializes as follows:

"Since Franklin D. Roosevelt took office as President of the United States in 1933 the Administration has waged a bitter war against big business and all that it stands for. The Government spokesmen used every means at their disposal to discredit big business. The latter was pictured as a gigantic, all-enveloping octopus which stifled the traditional freedom of action of every American citizen, and this idea was rammed home time and time again in one guise or another. The inevitable result was a very difficult time for big business. Labour was convinced that it had been exploited for years and demanded that conditions should be improved, sometimes when the cost of such improvements would impose an intolerable burden on individual companies. During the early days of this war, big business was virtually defenseless and the barbed shafts of criticism were sent on their way with the knowledge that damage could be done without fear of reprisal. However, the need for counter propaganda was soon recognized and the leading companies and associations began to equip themselves with men styled as Public Relations Counsels or Directors of Public Relations. It was the task of these men to ensure that the reputation of their employers should not be sullied and this they did by a mixture of whitewashing of big business and violent attack on its detractors. The methods employed were many, varying from lobbying at Washington to press and radio campaigns."

Metallurgical industries are of all sizes, but lots of them are big, and these are located in the front-line trenches in this undeclared war. The English editorial was confined to the name-calling and counter name-calling gas attacks of the war of "public relations." Gas is irritating, but only temporarily effective.

Among the high explosives that put the adversary permanently out of action are those developments that the man in the street can appreciate as counting heavily for one side or the other. After all, it is the man in the street who finally issues the "mandate" of the ballot and decides who won the war. For example, anyone who started driving a car when a $35.00 tire went but 3000 miles, has a strong conviction that tire makers have done a swell job. Individual tire companies may commit a lot of sins, but the whole industry is not on the defensive in the mind of the public at large.

The technical achievements of any industry are news today. Unproven sales claims, that attach a scientific-sounding name to some individual product or some kink in a process purely for advertising pur-

(Continued on page 134)

Carbon Linings for Blast Furnaces

by **FELIX SINGER**

Consulting Ceramist, 46 Castlemaine Ave., South Croydon, Surrey, England

Tapping Pig Iron from a Blast Furnace. (Courtesy: Ford Motor Co.)

Some months ago we were told that METALS AND ALLOYS *could do a service to the blast furnace industry of this country if we could get some authoritative information on the use of carbon linings for blast furnaces as used in Germany. It was stated that there is keen interest here in this subject.*

We were recommended to solicit a discussion from the author of this article, who formerly lived in Germany. He is a Dr.-Phil., Dr.-Ing. and a F.A.C.S. and ranks high as a specialist on refractories. Dr. Singer presents a discussion of the use of carbon linings which embraces Russia and Sweden as well as Germany.—The Editors

THE GENERAL REARMAMENT, and even more the potential readiness for war, of all industrial states and the production of steel which is thereby increased, all contribute to a growth of interest in refractory materials of every kind. Of particular interest are, of course, the refractory products for the manufacture of iron and steel.[1] New products such as olivine (forsterite) receive general attention. Products which have been used in few countries only, such as carbon bricks or carbon ramming masses, are now studied in many other countries.

In the United States of America, so far as I am aware, exclusive use is made of refractory bricks rich in alumina for refractory linings of blast furnaces. By the development of modern shaping processes, by carefully choosing high-class raw materials, and by burning at high temperatures, the quality of these bricks has essentially improved in the course of the last few years.[1a] Characteristic of the progress attained in this direction is the fact that about 15 yrs. ago with one blast furnace lining an average of 500,000 tons of pig iron was smelted, whereas at present the output usually exceeds 1,000,000 tons per lining, and in some cases has attained even the 2,000,000-ton limit.[2]

In Europe, too, blast furnace lining is built, to a great part, of refractory fire bricks of so-called "blast furnace quality." Besides that type of refractory building materials, however, for those parts of the lining—base, hearth and bosh—standing the highest requirements, refractory carbon bricks[2a] have been used for decades, and recently also unburned carbon ramming masses. Originally the use of carbon bricks in blast furnaces was practically limited to the German industrial area of the Lower Rhine.[3] Recently, interest in this material seems to have grown more general in Germany; a big exception just lately does not belie this, since the non-German builder of certain German blast furnaces had no practical experience with carbon linings but only with good refractory fireclay linings. Probably under the influence of German experts, carbon linings seem to be getting introduced into the big Soviet Russian iron industry which is now being built up.[4] Likewise they are finding special and interesting application in Sweden.

In spite of its lower apparent density, a lining of carbon bricks is more expensive than one of fireclay bricks. In Germany, the price per ton amounts to about RM 105 for fireclay blocks for the sub-hearth, and to RM 240 for carbon blocks or bricks. If carbon bricks, nevertheless, are used particularly by large works on an increasing scale, this is due to the fact that their use in furnaces of high total capacity seems to offer certain technical advantages which compensate the difference in price. Therefore, a short report on what is known about the manufacture and the use of carbon bricks and carbon ramming masses for blast furnaces may be of interest even for those metallurgists who do not yet use this refractory material in their blast furnaces.

The Manufacture of Carbon Bricks

Raw materials for the manufacture of refractory carbon bricks are coke and tar. The coke should have a low content of ash, with mechanical strength, and low reactivity. These requirements are met by a good foundry coke; gas coke and blast furnace coke are unsuitable, as bricks of gas coke are of low mechanical strength while blast furnace coke is highly combustible. The coke is, prior to its working up, dried thoroughly and crushed to a grain of about 1 mm (1/25 in.), in an edge runner mill, tube mill or ball mill, if necessary after having passed a jaw breaker. As a tar, a steel works tar free from water is used, which is a solution of pitch in the highest boiling constituents of pit coal tar. Frequently low proportions (about 10%) of hard petroleum asphalt, melting at a high temperature, are added to the tar. According to Russian reports, mixtures of tar, anthracene oil and bitumen[5] are recommended, to which small proportions of naphthalene may be added.[6]

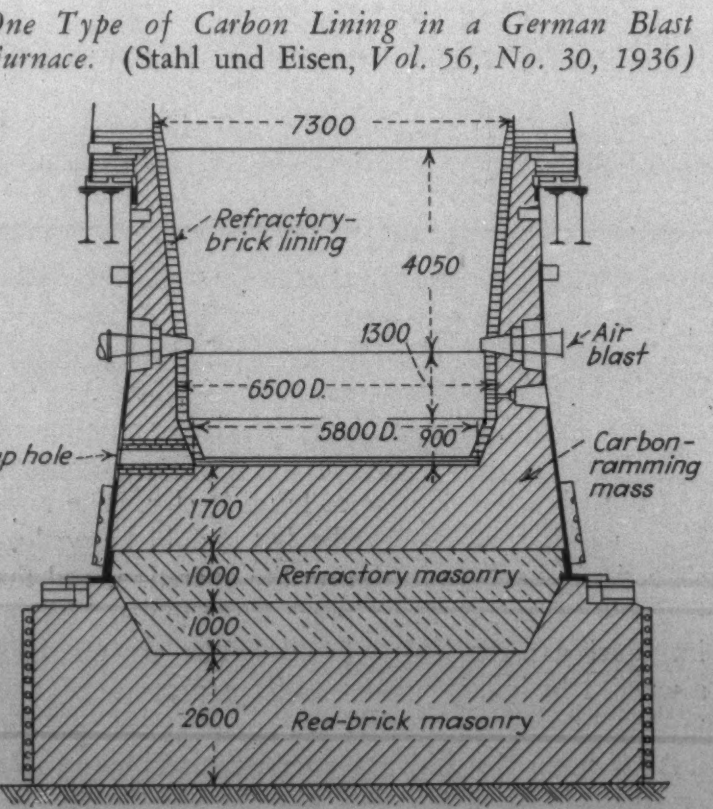
In any case the bond should contain as high as possible a proportion of such substances rich in free carbon which, when coked, yield a dense, mechanically strong, residue. The coke-tar mixtures contain 1 part in weight of tar to about 4 to 5 parts in weight of coke.

Prior to working up, the tar and coke are heated —mostly in steam heated containers—and mixed by hand or in kneading edge runner mills. A very thorough mixing is obtained by means of heated mixers of the Baker-Perkins type. The hot mixture is rammed into moulds either by hand or by means of pneumatic rammers, the same way as refractory bodies with a low content of binding clay. The rammers, which must not be flat but should be wedge-shaped, are kept warm during the operation, and the individual layers of the mixture are roughened before introducing more body. Small bricks are moulded also with hydraulic presses under a pressure of 200 to 400 kg. per sq. cm. (2850 to 5700 lbs. per sq. in.); it seems that hitherto no attempt has been made also to manufacture in this way the big blocks used in blast furnaces. The body rammed into the moulds whereby it solidifies. Even very big blocks may be put into a furnace 5 to 6 days after having been moulded.

Carbon bricks are burned in periodic single-chamber down-draught kilns, or in gas chamber kilns. In both cases they are put edgewise into dense muffles built of firebrick slabs, all interstices between the individual bricks and between the bricks and the muffle wall being carefully filled up with crushed coke. The uppermost layer of bricks is covered with a thick coke layer (about 20 cm = 8 in. thick) on which a flat layer of fire bricks is carefully embedded in refractory mortar or fireclay. Carlier[7] describes a special kiln having a removable roof and permanently built-in muffle walls, into which the bricks are set from above. The carbon bricks are burned mostly at a temperature of about 1000 deg. C. In order to attain this temperature inside the muffles built into periodic kilns and filled with bricks and coke, Cone No. 10 must, as a rule, come down in the kiln. In the periodic kiln, the bricks are preheated for about two days in order to let the volatile constituents of the tar escape slowly; then within two further days, the maturing temperature is attained, and then the kiln is cooled down the same way as a kiln set with fire bricks. The burned bricks are so ground that they can be set with close joints.

Carbon ramming masses[8] are produced the same way as carbon bricks and from the same raw materials. They contain about 6 parts in weight of coke to 1 part in weight of tar. These masses, too, are supplied hot when constructing the monolithic lining of the sub-hearth, and are rammed by hand in thin layers (about 5 cm = 2 in. thick) by means of hot iron rammers.

One Type of Carbon Lining in a German Blast Furnace. (Stahl und Eisen, *Vol. 56, No. 30, 1936)*



Properties of Carbon Bricks for Blast Furnaces

Carbon bricks react neither with the slags formed in the blast furnace nor with the alloy constituents of the pig iron. The opinion that carbon bricks are destroyed in the blast furnace by carbon being given off in the melting, is wrong.[9] Pig iron is saturated with carbon in the blast furnace and therefore cannot absorb larger quantities of carbon.

Carbon bricks fuse at temperatures far above Cone No. 42. Exact data as to the softening under pressure of good blast furnace carbon bricks are not yet available, but their resistance is much better than that of the best quality fire bricks.

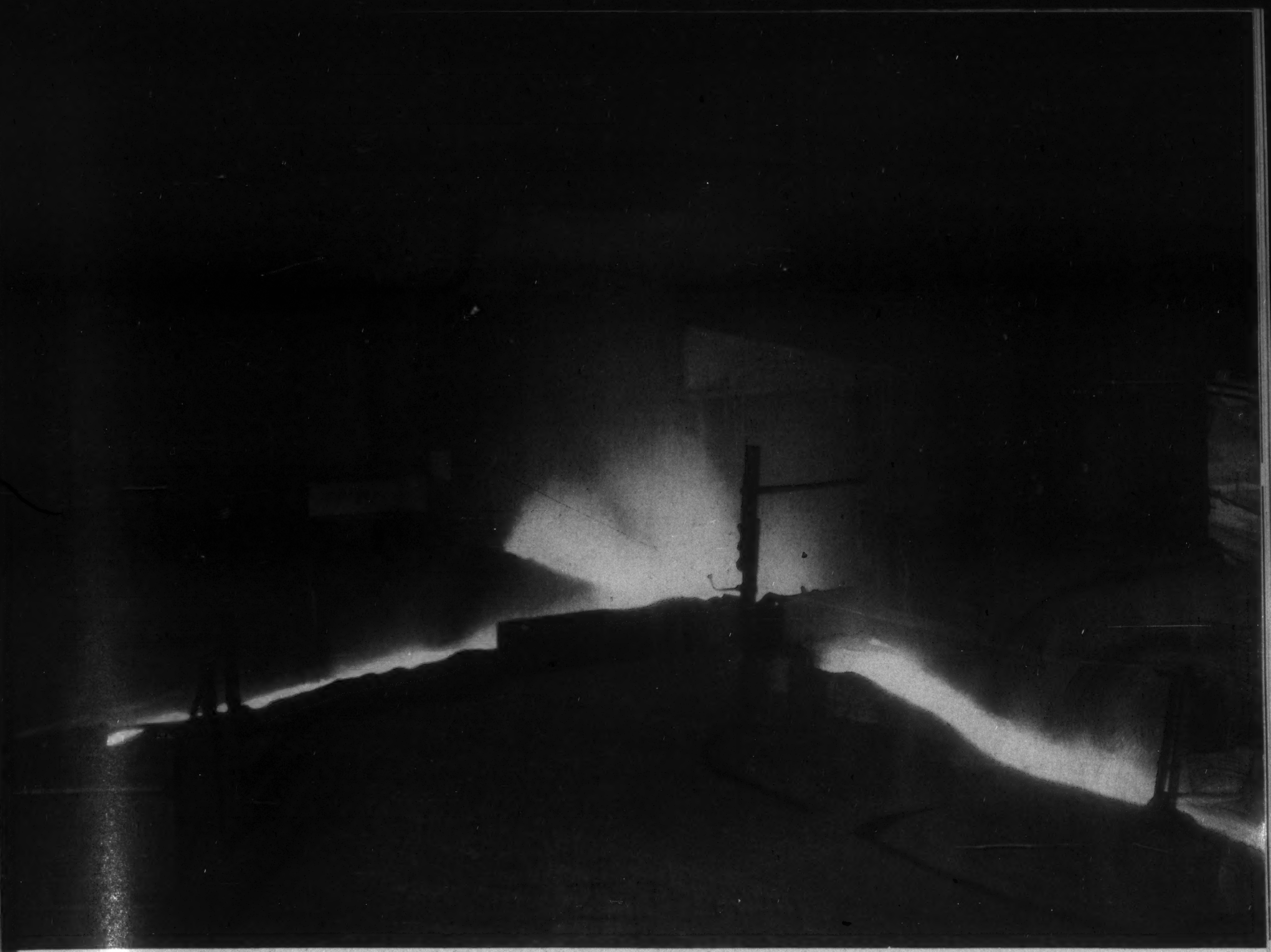
The heat conductivity of carbon bricks is 15 to 20 times that of fireclay bricks. The coefficient of expansion of the material within the range 0 to 900 deg. C. is about 5.8 x 10^{-6}, the expansion curve being practically a straight line.

Carbon blocks should be appropriately dimensioned, have smooth faces and sharp edges, and must not contain any porous spots. Their content of ash must not exceed 10 per cent, their content of carbon should amount to at least between 85 and 90 per cent, and their compression strength in the cold state should be 120 kg. per sq. cm. (1710 lbs. per sq. in.) or more. German carbon bricks at present have compression strengths of 180 to 300 kg. per sq. cm. (2600 to 4300 lbs. per sq. in.), a water absorption of about 21 to 23 per cent, a total porosity of 30 to 32 per cent, and an apparent specific gravity of about 1.35. Approximately equal values to these properties have been measured with bricks of Russian make.

Carbon Bricks for Blast Furnace Linings

Since carbon bricks are unburnable, only in a reducing atmosphere, linings made of burned bricks or rammed masses must be protected against destruction from the starting up of the furnace until the sub-hearth has been filled with iron and slags. The lining is therefore, after having been built in, covered with a layer of fire brick slabs embedded in a refractory mortar which, although soon destroyed by the iron and slags, are sufficient to protect the lining as long as there is any danger of oxidation. Carbon bricks cannot be built in immediately adjacent the tap hole, as they would burn out there. Even when the base, hearth and bosh consist of carbon blocks, the parts surrounding the tap hole of the lining are masoned in fireclay blocks.

Blast furnaces lined with carbon bricks and kept running for 5 to 6 yrs., attained an output of 1,000,-



Pig Iron Being Tapped from a Blast Furnace. (Courtesy: Youngstown Sheet & Tube Co.)

000 tons of pig iron per life of the blast furnace.[10] One furnace of the Dortmund-Hoerder Huetten Verein, the sub-hearth of which was not masoned of blocks but made of carbon ramming masses, and which was put into operation at the end of 1929, was still working in 1936. With a total of 1129 days of stoppage, altogether 830,000 tons of pig iron were produced in this furnace up to 1936.[11] By then, eight furnaces were put into operation in Germany, all having linings of carbon masses. The number of furnaces in which carbon bricks are used, is considerably greater, but the actual figure is obtainable neither in Germany nor Russia.

Carbon bricks have higher refractoriness, and resist softening under load better than fireclay bricks, and—in contrast to fireclay bricks—they cannot react with any constituents of the blast furnace slags on account of their chemical nature. Carbon bricks may be produced without difficulty in the form of very big blocks. The use of such blocks is an advantage, since the number and length of the joints in the linings can be considerably reduced. While, for example, in America fireclay blocks in dimensions of 450 x 225 x 120 mm. (about 16 x 9 x 4¾ in.) and in Russia[12] smaller blocks of 345 x 150 x 175 mm. (about 13½ x 6 x 7 in.) are used for the base, carbon blocks in the dimensions of 700 x 500 x 250 mm. (about 17¾ x 12¾ x 6⅜ in.) are usual at present in Germany, but pieces up to a cubic capacity of about 0.5 cm.^{3} (17½ cu. ft.) can be produced. As the carbon bricks do not react with the charge, the danger of deposits of refined iron and of "baking-on" (i.e. the formation of so-called blast furnace "sows"), is reduced. On account of the high heat conductivity of the carbon bricks and also because these bricks can absorb large quantities of cooling water due to their high porosity, the cooling of a lining consisting of such bricks is much more intense than that of a lining consisting of fire bricks. This circumstance can be very important for the lifetime of the lining; especially so if the lining is very much damaged in places, for, by intensive cooling, the danger of a breach can be combatted much better than in the case of fireclay brick linings.

Two of the 12 Blast Furnaces at Gary Works of the Carnegie-Illinois Steel Corp. with the Hot Blast Stoves between them.

Recent figures published as to the efficiency of American and German blast furnaces do not permit a clear judgment whether carbon bricks as a lining of the sub-hearth really have a longer life than the best quality fire bricks as now used in America. The efficiency of a blast furnace lining does not only depend upon the properties of the refractory material, but also on the charge, the design of the furnace, the duration and number of stoppages and a large number of other factors. Therefore the figures of different furnaces cannot be directly compared.

From a great number of figures one may, however, conclude that American linings which entirely consist of fireclay bricks have no shorter lifetime, or at least not a considerably shorter one, than those German linings of which the sub-hearth consists of carbon bricks or carbon ramming masses. One may, however, assume that the German high quality blast furnace bricks of the aluminum silicate type are not inferior as regards technical properties to the material at present on the market in America. [This is a natural attitude for the Germans to take. However, the lack of flint clays in Germany has led some observers to believe that the quality of German blast furnace fireclay brick is not on a par with that of American products utilizing flint clays in the body. —Editors] If, nevertheless, there has been recently a tendency in Germany (with the above mentioned exception) to use carbon bricks for high efficiency furnaces on a larger scale than before, this seems due to the fact that carbon bricks give greater reliability with equal efficiency to a furnace on account of the possibility to build them in with less joints and to cool them more intensely as well as on account of their chemical nature which excludes any reactions with the charge, slags or metal. Though higher efficiency of the carbon bricks cannot be proved as expressed in tons of pig iron per lining only, or if one compares the figures of the best quality fireclay bricks with those of normal carbon bricks, the more expensive refractory carbon products might be, nevertheless, on account of their higher reliability, a more economical furnace building material than the best fireclay bricks at present produced in Germany or in America. Therefore it was the aim of this report to draw, in rough outlines only, attention to a material of which the application has offered certain advantages not only for Swedish, German and Russian, but also for American iron works, which must not be under-estimated, since, as far as I am aware, no indications regarding production and availability are to be found in American literature.

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See Comments on Page 133

Properties of Some Free-Machining, Lead-Bearing Steels - Part 2

by J. H. NEAD, C. E. SIMS and O. E. HARDER

Mr. Nead is chief metallurgist, Inland Steel Co., Chicago, and Messrs. Sims and Harder are of the staff of Battelle Memorial Institute, Columbus, Ohio.

We present herewith the second and last installment of this discussion of the properties of "Ledloy," the lead-bearing open-hearth steel which has remarkable free-machining properties. In the first section, published in our March issue, the following points were discussed. Distribution of the Lead; Is the Lead in Solution?; and Mechanical Properties. An article on the manufacture of Ledloy is expected to be forthcoming.—The Editors.

Hardenability

Hardenability curves for steels of the higher carbon group whose compositions have been given above were determined on quenched 7/8 in. dia. specimens after tempering at 300 deg. F. The results are shown in Fig. 11.

At these quenching temperatures, there is no appreciable difference in hardenability. Under heating conditions where an adherent scale is formed and quenching is therefore slower, the steels with lead are more shallow hardening by a very slight amount but, in the group of steels, this effect becomes plainly evident only in the 1045. Robbins shows a slight decrease in depth hardening of 1-3/16 in. rounds of water quenched steel 1045 containing lead, with the surface hardness the same as in the companion steel without lead.

A comparison of the "core properties" i.e. the properties of uncarburized regular and lead-containing S.A.E. 1115, put through the carburizing time, temperature and heat-treatment cycle, gave the following:

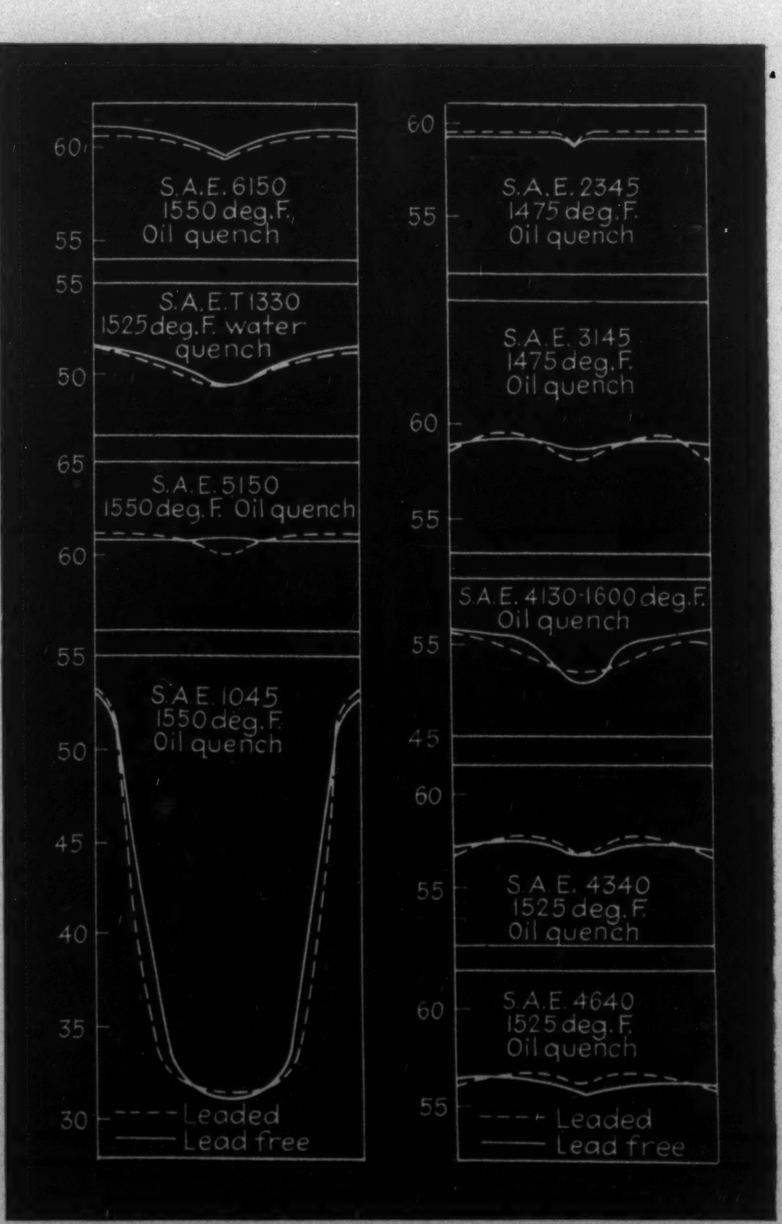


Fig. 11. Hardenability Tests on High Carbon Alloy Steels Leaded and Lead-Free. Specimens are 7/8 in. diam.

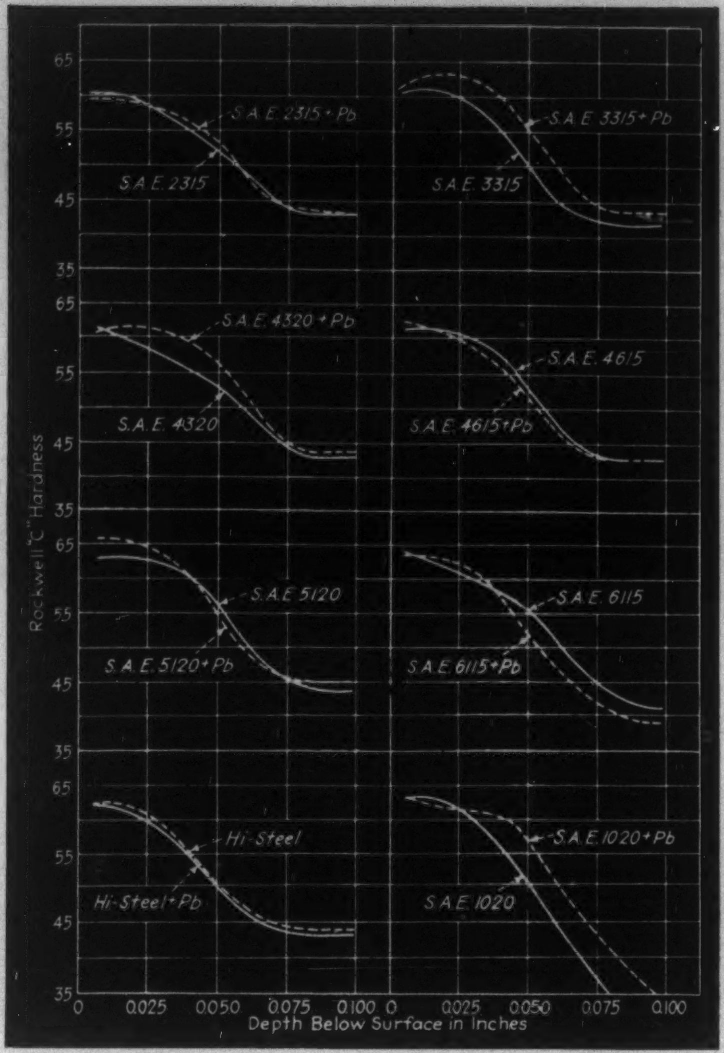


Fig. 12. Hardness Tests on Carburized and Quenched Samples.

	Single Quench		Double Quench	
	Regular 1115	Ledloy* 1115	Regular 1115	Ledloy* 1115
Yield strength, lbs. per sq. in.	54,500	59,500	49,500	48,500
Ultimate strength, lbs. per sq. in.	85,000	82,500	75,500	73,500
Elongation in 2 in., per cent	23	22	28.5	31
Reduction in area, per cent	61	62	60.5	66
Brinell hardness	163	163	156	149
Charpy impact, ft.-lbs.	38	32	42	43

* 0.20 to 0.30% Lead.

Carburizing

The lower carbon group of steels in Table IV was carburized, as ½ x ½ x 2½ in. specimens, for 8 hrs. at 1700 deg. in compound. The specimens were water quenched from the pot and taper ground for comparison of hardness penetration. The curves are shown in Fig. 12. The lead contents of the lead-containing specimens were:

Type of Steel	Pb Content	Type of Steel	Pb Content
1020	0.19	4615	0.18
2315	0.17	5120	0.12
4320	0.11	Hi-Steel	0.18
3115	0.14	6115	0.16

The tests cited above and many others not reported were made with the aim of a general survey to detect any noteworthy trend. From such tests only one conclusion can be drawn, i.e., that no noteworthy trend exists, since any given property scarcely varies more than is customary in duplicate tests on non-leaded steel.

It appears that the presence of a soft material like lead, in these small quantities and in this remarkably fine and uniform state of dispersion, does not notably depreciate the mechanical properties, so that the gain in machinability is obtained without material sacrifice. It is concluded that the addition of lead in proper percentage and distribution to the steels studied does not markedly alter their properties, other than machinability.

Machinability

The early development of the lead-bearing steels was especially aimed at screw stock. Many experimenters have concluded that it is a very difficult task to obtain a true indication of differences in machinability by any laboratory test, and that little of value can be found out without recourse to production tests on automatic screw machines. Laboratory tests were therefore undertaken with a good deal of skepticism and with the fear that they would not be dependable, though they might give an indication of trends.

It was pleasantly surprising to find that not only did laboratory tests show a decided difference between lead-free and lead-bearing stock, but that the direction of these differences and their magnitude were quite closely paralleled by production tests on automatic machines.

Two laboratory tests for machinability were employed, sawing and drilling, both being designed to show comparative values based on an arbitrary standard. The saw test was carried out on a power hack saw using new saw blades of a good quality. The stock was in the form of 1-in. rounds. The saw operated at constant speed and was allowed to feed freely as fast as it could cut.

In carrying out the test the saw blade was standardized by making five cuts through an arbitrary standard, which in this case was a stock of 1-in. round S.A.E. 1020 cold-drawn steel. Then five cuts were made through the bar under test. The time in seconds to cut through the bar was measured with a stop watch. To minimize the influence of a slight dulling of the saw blade three cuts were then made through the standard bar followed by five cuts through the test bar, and then two more cuts through the standard. The test was then repeated using another saw blade.

By dividing the average time required to cut through the standard by the average time for the test bar, a ratio or index was obtained which was designated the sawability index. For instance, if it required 90 secs. to saw through the standard and 80 secs. for the test steel, the latter would have a sawability index of 90x100/80 or 112.

The drill test was made on a standard type of drill press with some special attachments. The counter weight for the spindle was disconnected and in place of the hand feed levers a circumferentially grooved disk was substituted. A light cable was wrapped in the groove and a weight suspended on one end. This means provided a constant end thrust of 84 lbs. to the drill which was run at a constant speed of 860 r.p.m. The drills were ¼ in. high speed twist drills, sharpened in a jig.

In making the tests the drill was allowed to penetrate to its full diameter and then timed while it advanced a measured inch. By comparing the time required to penetrate 1 in. in the standard steel to the time required in the test steel, an index called the drillability index was obtained.

Regardless of how carefully the drills were ground, no two drills gave identical results and it was found advisable to drill at least three holes in the standard and each test steel with each drill and to use at least five different drills. The indices were then obtained by comparison of the grand average of all the tests.

A combined machinability index was obtained by taking the mean of the two types of test. Typical results of these tests on commercial steels, obtained through commercial channels, are found in Table IX.

Table IX—Machinability Indices Obtained on Typical Commercial Screw Stock Steels

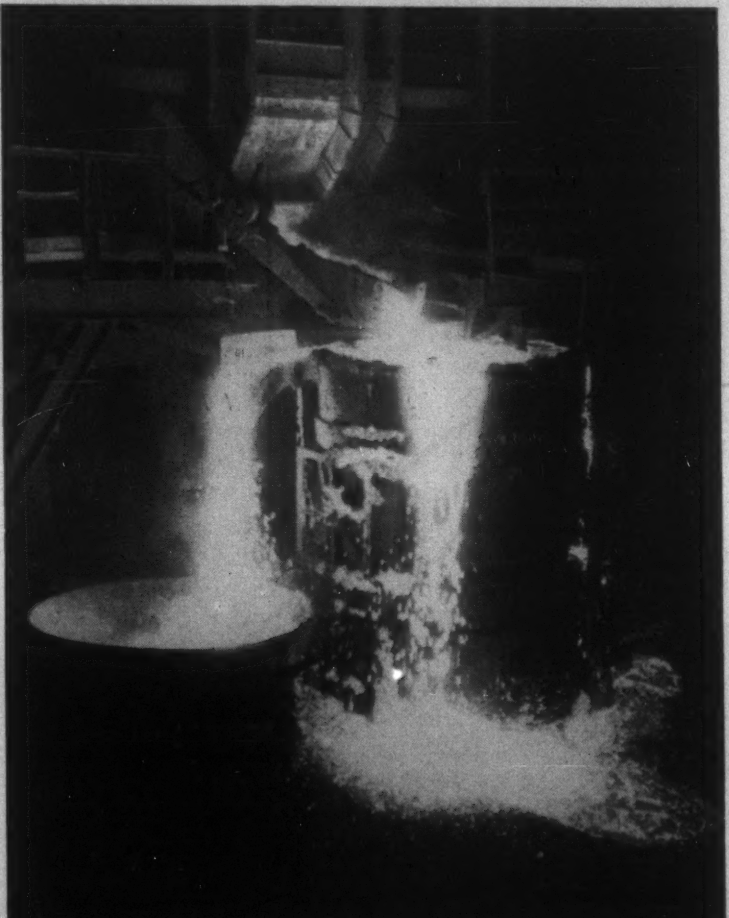
Type of Steel	Sawability index	Drillability index	Combined index
S.A.E. 1020 *	100	100	100
S.A.E. 1120	114	105	110
S.A.E. 1112	130	109	120
S.A.E. X1112	139	116	128

* Arbitrary standard.

Six lots of S.A.E. 1112 Bessemer screw stock, supposedly of equal grade, were tested for machinability. The indices obtained on three of these were 116, 120, and 124. The other three lots each had an index of 127. Tests on the steels of Table I gave the machinability data of Table XI.

Although there is no belief, and no intent to imply that the tests described constitute a complete and thorough test of machinability, they, nevertheless, consistently showed differences in steels. Subsequent tests on automatic screw machines proved these differences to be of significance as regards machinability.

From the data available at this time, the indications are that the same order of machinability as determined by the laboratory tests has always been found in commercial tests. Commercial tests, how-



Topping an Open-Hearth Furnace Showing Slag Overflow from Top of Ladle. (Courtesy: Inland Steel Co.)

Pouring Open-Hearth Steel from the Ladle into Ingot Molds. (Courtesy: Inland Steel Co.)

ever, show greater differences than are shown by the laboratory tests. This is evident in that a Bessemer screw stock (X1112) will actually machine 50 to 100 per cent faster than a 1020 steel though the difference shown by the laboratory tests is only 28 per cent. Fig. 13 shows both commercial and experimental high sulphur steels with varying lead content, plotted against the combined machinability index.

The rather spectacular improvement produced by lead in the production machining of screw stock has been clearly set forth in the article by Robbins. Some production data are summarized below:

Lead-containing open-hearth S.A.E. 1120 was put on machines running on high sulphur Bessemer S.A.E. X1112 (C—0.08-0.16; Mn—0.60-0.90; P—0.09-0.13; S—0.20-0.30 per cent). In practically every case it was possible to increase both speed and feed on the machine after changing to the lead containing steel. The results of typical tests are shown in Table X. When it is considered that X1112 is a steel in which all other properties have been relegated to secondary importance in favor of machinability, the results of these tests are quite remarkable.

Table X—Commercial Machining Tests on Steels With and Without Lead

Machining Operations	Type of Steel	Surface speed, ft. per min.	Feed in inches per rev.	Time per piece, sec.	Increased production by lead-alloyed 1120, per cent
Drill Recess Cut off	S.A.E. X-1112	215	0.0102	9.0	
	Lead Alloyed 1120	250	0.0138	7.2	25
Form Thread Drill Ream Chamfer	S.A.E. X-1112	240	0.0038	43	
	Lead Alloyed 1120	366	0.0051	25	72
Form Drill Knurl Chamfer	S.A.E. X-1112	220	0.0022	9.0	
	Lead Alloyed 1120	220	0.0045	6.7	34

The data in Table X show in the first and third tests an increased rate of production resulting principally from increased feed, but the second test shows what is possible by increasing both cutting speeds and feeds. In this test the surface cutting speed was increased over 50 per cent and the feed by 35 per cent, and yet, with the steel containing lead, the tools were run for 13 hrs. with regrinding. With the X-1112 steel the tools burned and production could not be maintained when the increased speeds and feeds were used.

Other comparisons in production machining operations were:

Part	Operations		S.A.E. 1020	Ledloy 1115 *
3⅛-in. Washer	Turning	Time per part	17 min.	5 min.
			S.A.E. X1315	Ledloy X1020 *
Small Gear Blanks	4 Spindle Automatic	Spindle speed	332	407
		Pieces per hour	76	112
			S.A.E. 1020	Ledloy X1020 *
Gear Blanks	Automatic Screw Machine	Parts per hour	92	150
			S.A.E. X1112	Ledloy 1115 *
Bushing Threaded Inside and Out	Automatic Screw Machine	Parts per hour	654	838

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* 0.20 to 0.30% Lead.

As soon as it became evident that increased machinability with no notable effect on mechanical properties was produced by lead in screw stock, the question arose whether an analogous situation exists in low sulphur steels and in strong, heat-treated steels. Getting a complete answer in the cases of all the S.A.E. and other commercial steels is obviously a long task, requiring years of experiment and observation of commercial practice. The initial tests, however, indicate that the possibilities are no less interesting. Some examples of tests on steels of higher hardness are given in Table XII.

For comparison of machinability of heat-treated steels at a hardness level far above that of screw stock, some other base of comparison than screw stock would be required. Instead of attempting to pick a single steel and treatment as the basis for comparison, the preliminary tests have been made on the basis of comparison between lead-free and lead-bearing steels from split heats.

Alloy steels taken from the series of split heats whose composition is given in Table IV were studied. To make the test a searching one, a high hardness level was chosen, the steels being quenched and drawn back to 350 Brinell.

For these drilling tests the specimens were quenched and tempered 1-in. rounds which had been heat-treated as follows:

S.A.E.	(½ Hr. at Temp.) Quenching Temp., deg. F.	(2 Hr.) Draw Temp., deg. F.
2345	1475	975
4130	1600	1000
T1330	1525	900
4340	1525	1125
3145	1475	1000
4640	1525	1000
5150	1550	1075
6150	1550	1125

The test consisted of drilling a series of ¼-in. holes, ½ in. deep, in each steel and observing the rate of dulling of the high speed drill. The rate of dulling was measured by the increased time consumed in drilling consecutive holes. In conducting the tests, three freshly sharpened drills were used on each lead-free steel. After completing the tests on the lead-free steel, the same three drills were resharpened on an automatic drill grinder, so as to duplicate the original conditions as closely as possible. The resharpened drills were then used in obtaining comparable "drill life" data on the lead-bearing steels. Spindle speeds of 1260 r.p.m. and a constant thrust of 84 lb. were used in these tests. Average results of the three drill tests are shown in Fig. 14.

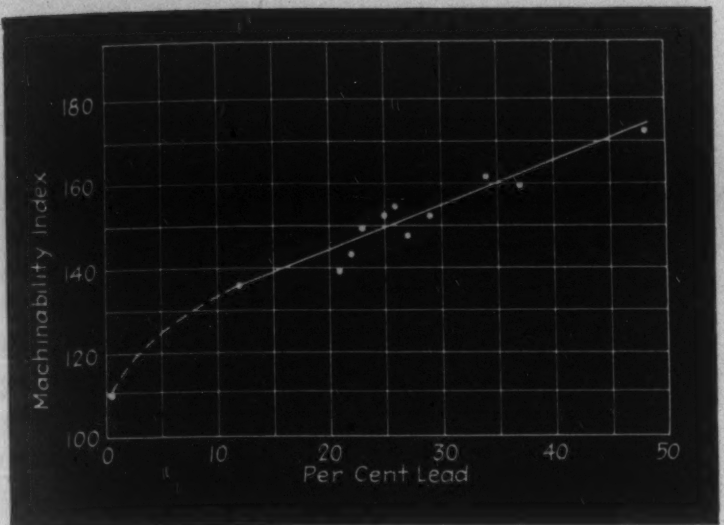

Fig. 13. Relation between Lead Contents and Average of Sawing and Drilling Machinability Indices.

Speculations on the Mechanism of Machinability Improvement

The usual discussions of machinability point out that improvements in machinability may be secured by altering the microstructure, i.e., producing a different distribution of ferrite and cementite plates (in pearlite) or of cementite spheroids in spheroidized material; by coarsening the grain; by increasing the brittleness as by materially increasing the phosphorus or nitrogen content of ferrite; by introducing nonmetallic inclusions, sulphides, selenides and the like to supply chip-breakers; by alloying additions that increase the rate of work-hardening and thus embrittle the chip; and by alloying additions that increase the reduction of area, a phenomenon thought to accompany localization of stress at the apex of the cut.

The effect of lead seems to fall into none of these categories and to be additive to the effects produced by factors that do fall into those categories. Lead does not appreciably affect the rate of work hardening, as is evidenced by the same hardness and ductility on cold-drawn lead-free and lead-containing stock. Lead does not alter the microstructure; its effect on austenitic grain size is very mildly in the

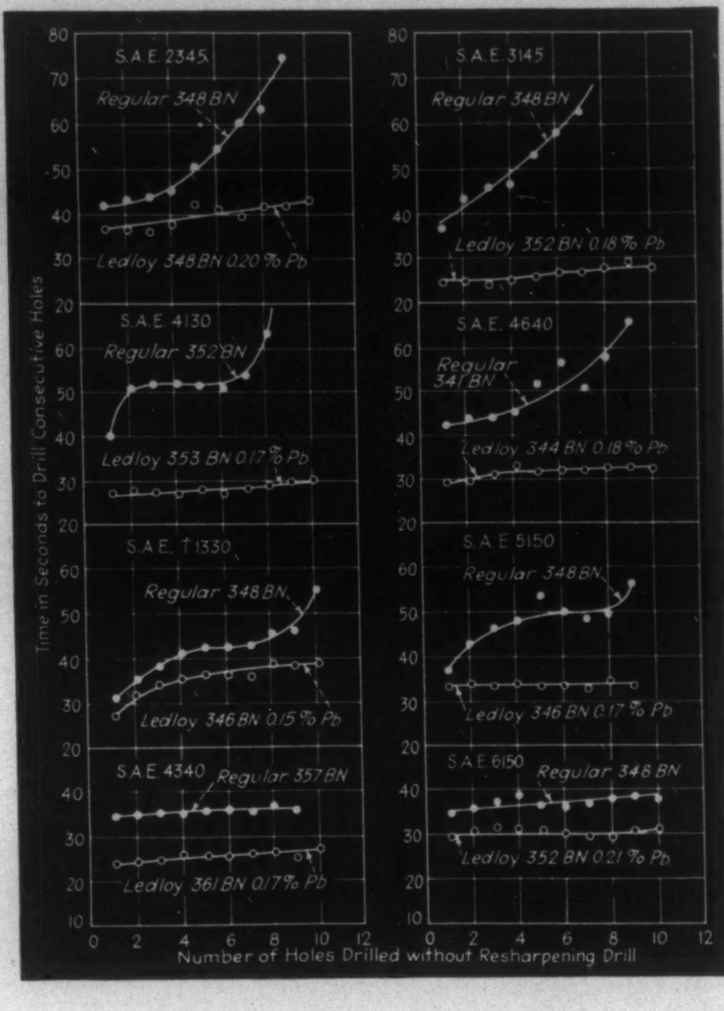

Fig. 14. "Drill Life" Tests.

Table XI—Machinability Data on Steels of Table I

Heat No.	Composition, Per Cent						Machinability Index		
	C	Mn	Si	P	S	Pb	Sawing	Drilling	Combined
2962	.11	.63	.012	.017	.19	..	119	102	111
2966	.10	.55	.012	.019	.20	.12	145	125	135
2967	.11	.59	.010	.017	.21	.26	172	137	155
2968	.11	.58	.010	.019	.21	.34	189	143	166
2969	.11	.62	.008	.017	.23	.48	212	145	179
2965	.12	.57	.014	.018	.10	.09	123	112	118
2974	.14	.69	.248	.019	.20	..	123	114	119
2963	.11	.68	.238	.020	.21	.09	159	122	141
2973	.11	1.19	.228	.020	.19	..	130	114	122
2975	.12	1.20	.203	.020	.20	.09	164	141	153
2964	.15	1.11	.010	.020	.20	.09	161	125	143
2991	.10	.29	.104	.016	.26	.39	200	147	174

See Table I for mechanical properties.

Table XII—Composition and Sawability Indices of Experimental Low-Sulphur Steels Showing Effect of Lead

Heat No.	C	Mn	P	S	Si	Cr	Ni	Mo	Pb	Heat Treatment	Brinell Hardness	Sawability Index
3494	.15	.54	.024	.025	.09	..	...	..	..	Norm. 1600 deg. F................	114	107
3495	.17	.85	.025	.025	.11	..	...	..	.07	Norm. 1600 deg. F................	121	137
3496	.47	.74	.027	.025	.09	..	...	..	..	Norm. 1500 deg. F................	179	147
3497	.46	.80	.024	.025	.17	..	...	..	.20	Norm. 1500 deg. F................	179	196
3498	.88	.74	.022	.024	.16	..	...	..	..	Norm. 1450 deg. F................	269	139
3499	.88	.82	.023	.025	.15	..	...	..	.18	Norm. 1450 deg. F................	277	179
3502	.48	.74	.022	.017	.14	.72	1.42	.16	..	Oil Quench from 1500 deg. F. Draw 1000 deg. F.................	341	137
3503	.49	.77	.024	.015	.15	.75	1.84	.17	.16	Oil Quench from 1500 deg. F. Draw 1000 deg. F.................	341	173
3502	.48	.74	.022	.017	.14	.72	1.42	.16	..	Annealed from 1500 deg. F. Draw 1250 deg. F.................	210	156
3503	.49	.77	.024	.015	.15	.75	1.84	.17	.16	Annealed from 1500 deg. F. Draw 1250 deg. F.................	210	182

direction of finer grain, which, save for smoothness of finish, would be in the wrong direction; it does not increase impact brittleness, even at temperatures up to 600 deg., by anywhere near the amount one would expect to be required for differences in machinability of the magnitude noted. It does not enhance reduction of area, and it gives no microscopically visible chip-breaker pools of lead so that there is no close similarity to the effect of lead in leaded brass. The chips from lead-containing soft steels are shorter than with comparison steels free from lead and come off with a yellow temper color film under cutting conditions where the chips from the comparison steel come off blue. Not much difference in appearance of chips was noted in the tests on steels of 350 Brinell. While somewhat better chip-breaking effects may be responsible for some of the improvement, the lower chip temperature seems the more significant in respect to the mechanisms involved.

One might expect a certain degree of "soapiness" from the analogy of the ability of lead to reduce friction. In referring to the mechanism of the effect of manganese sulphide inclusions, it has been suggested (U.S.S. Carilloy Steels, 1938, p. 183) that "sulphide particles may possibly be smeared out and tend to act as a highly efficient lubricant which reduces the abrasion on the tool and aids in producing a smooth finish." Such a mechanism of the possible effect of submicroscopically distributed lead is fairly satisfying to the mind, and one readily sees that extremely uniform dissemination would facilitate such an action. Definite proof that the behavior is essentially that of lubrication is, however, lacking, though no observation seems to be inconsistent with such an assumption.

The only other steel in which a fine dispersion of soft, submicroscopic particles is postulated, is a precipitation-hardened copper steel. In that case the particles, even though softer than the matrix, notably raise its tensile and yield strengths, (a feature absent in the lead-bearing steels), drop elongation, reduction of area and impact resistance to a much more marked degree than in the case of lead, and produce no such improvement in machinability.

The effects of lead appear very specific, and the peculiar combination of results it produces could not have been anticipated.

Other Features

One without experience in the manufacture and fabrication of lead-bearing steels might anticipate more difficulties than actually exist in the steel plant, and that would exist were the lead distributed in pools, rather than in a state of submicscroscopic dissemination. The rolling of the lead-bearing steels proceeds quite exactly like that of the same steels without lead. As long as no lead pools are allowed to form, there are no losses by cracking during rolling traceable to the lead content.

All kinds of industrial hazards from lead fume in melting, pouring and heating could be visualized and might seem real obstacles. Actually, the development of proper sanitary precautions went hand in hand with the commercialization of "Ledloy" at Inland and this experience has been passed on to the licensees, so that the technique of operation without health hazard is accomplished. Furthermore, the hazard does not appear to be as great as in some other industries, for example, the manufacture of high lead bearing metals and lead refining where these problems have been long present and satisfactorily solved.

Some of those who do not operate steel melting furnaces are exercised by the fear of dire results to furnace refractories and on the properties of steels not intended to contain lead, when lead-bearing scrap finally comes back to the furnaces. However, those who use miscellaneous scrap, such as that from which high-lead bearings have not been removed, often find lead seeping through a furnace bottom, large quantities of lead sometimes being removed after a long campaign, without harmful evidence of the presence of lead.

Experience of Inland in the use of ingot crops from "Ledloy" does not indicate that these fears in respect to refractories will be realized, and the data on mechanical properties cited above should allay those in respect to "residual" lead from scrap. Doubtless a spectroscopic trace of lead will be seized upon as a convenient whipping-boy for the explanation of miscellaneous ills of steel, whatever they may be, just as has been the case with other residual metals, but in time such matters settle down to a basis of fact rather than speculation. Certainly the fear of poisoning of either men or steel is not held by those at Inland who have actual commercial experience with the use of lead in steel.

The beneficial use of lead in steel runs counter to metallurgical expectations, which would have predicted inability to get uniform dissemination because of reported liquid and solid insolubility. There is no background of long time experience; lead is but an infant in the kindergarten of the alloying elements. As regards screw stock, it is quite a lusty infant. As it grows and passes up through the grades, high school and college, it may be graduated into an important alloying element in a much wider field, since an improvement in machinability of strong, heat-treated steels, without introducing stress-raising inclusions, or deteriorating the mechanical properties, is a consummation greatly to be desired and one that seems to be within hailing distance.

The development of lead-bearing steels of improved machinability has been carried on jointly by the metallurgical department of the Inland Steel Co. and by workers at Battelle Memorial Institute, the work at Battelle being sponsored by Inland.

The authors are glad to acknowledge the valuable work done by J. W. Halley, T. S. Washburn and others of the Inland Steel Co., and of F. B. Dahle, Hugo Johnson and H. G. Holm of the Battelle staff.

A CRITICAL REVIEW • PART 2

Grain Size of Steel

by N. F. WARD and J. E. DORN

Associate Professor of Mechanical Engineering and Assistant Professor of Mechanical Engineering, Respectively, University of California, Berkeley, Cal. Prof. Dorn was formerly Research Associate at Battelle Memorial Inst., Columbus.

This instructive series—an interpretation, for the metallurgical engineer, of one of physical metallurgy's shining achievements—opened in our March issue with a discussion of the real importance of grain size of steel in the general industrial picture and a description of the best-known methods of testing austenitic grain size. Part II, published herewith, covers the various methods of reporting austenitic grain size and describes the widely-used and very practical fracture test, known in this country as the Shepherd P. F. test. The subject of the effect of grain size on hardenability and heat treating practice, to be discussed fully in Part III, is introduced in the present installment.—The Editors.

Methods of Reporting Grain Size

The grain size of metals and alloys had been of interest to metallurgical engineers and metallurgists for many years before the importance of austenitic grain size was recognized. Numerous methods, therefore, for reporting the grain size of metals had been developed. Each method appears to have certain advantages over other methods for a particular study. Since this discussion is limited to a study of austenitic grain size only, it will be unnecessary to describe all of the methods that have been developed and their special advantages in certain types of investigations. Practically all of the methods of actually measuring austenitic grain size are based on that outlined by Jeffries.[1]

Grains, in general, are three-dimensional polyhedral solids. The actual shape of the grains depends upon the number and distribution of the nuclei from which they arise. Each nucleus grows by the adsorption of untransformed metal until it meets another similarly growing grain. The two advancing grain fronts meet in a plane which thereupon assumes the significance of a grain boundary. This accounts for the polyhedral nature of the grains. The initial grains, formed in this manner, may continue to grow by the adsorption of some grains (generally the smaller) by others. Nevertheless, their shape remains practically polyhedral.

Microscopic observation of grain size is made by examination of a polished and etched plane section of a specimen at a certain magnification. The quantity noted, therefore, is not the actual grain volume but some average value of the cross-sectional area of an average grain. Jeffries[1] has outlined a simple method of recording this grain size. A circle of known area, A, is drawn on a photomicrograph of a representative section of the specimen at a known magnification, M. A ground glass screen may be used in place of a photomicrograph. Since the grain sizes may vary, not only because some grains are actually larger than others, but also because the polished plane may cut some grains in planes which truncate the grains at points where their cross-sectional area is small, Jeffries[1] recommends that the area of the circle, A, and the magnification, M, be adjusted so that at least fifty grains be included in the circle. For some work even this value may be too small. The sum of the number of grains totally enclosed in the circle plus one-half of the number of grains intercepted by the circumference of the circle may be reported as the number of grains per area A at a magnification M. Since it is customary to report the austenitic grain size as the number of grains per sq. in. at 100 diameters, it is advisable to transform the above determination to satisfy these requirements. If n′ is the number of grains reported for an area A

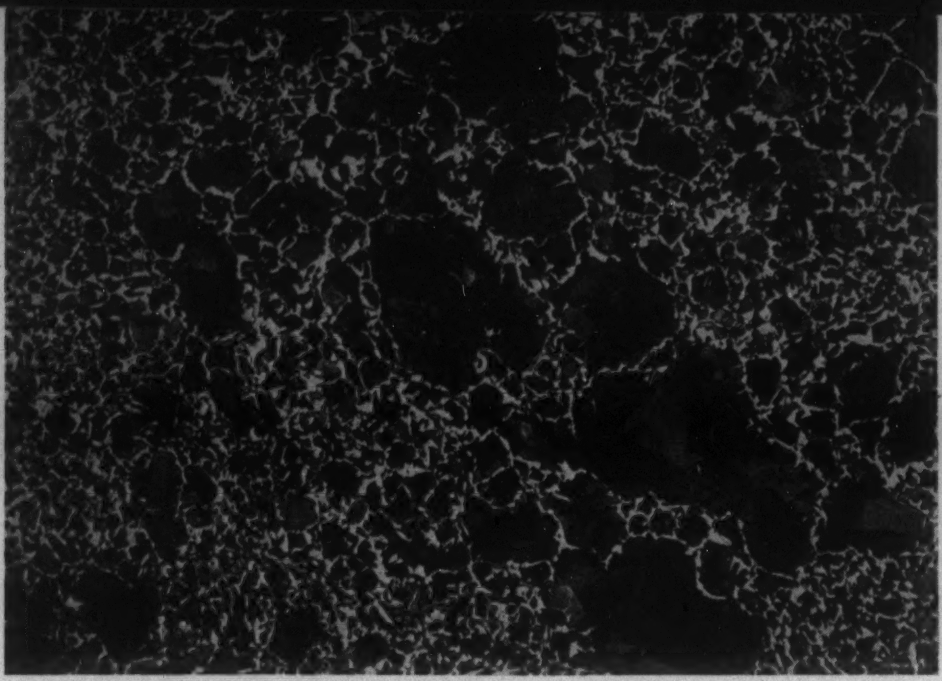

Fig. 8. Duplexed Austenitic Grain Size as Revealed by the Gradient Quench.

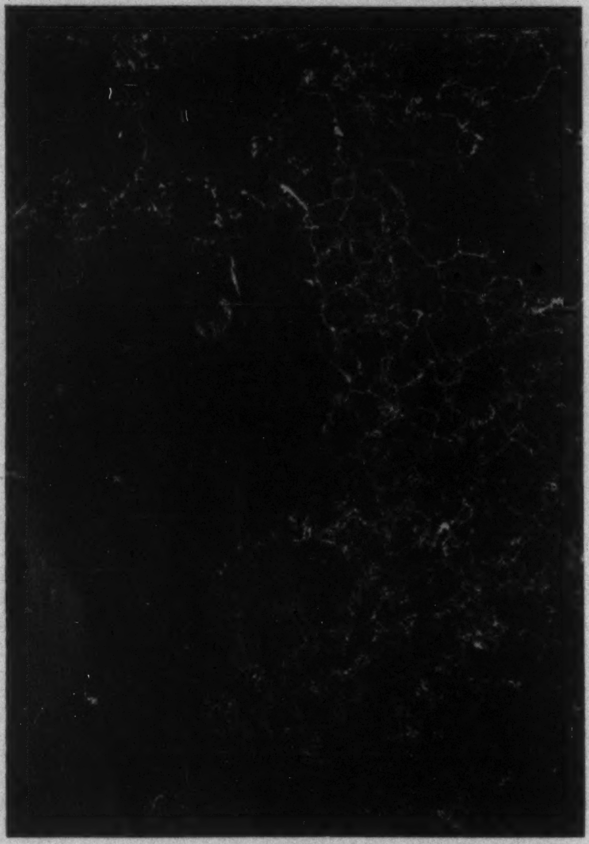

Fig. 9. Extremely Duplexed Grain Size Observed in Recent Investigations by Lauderdale and Harder on Case Carburized S. A. E. Steels. (Courtesy of O. E. Harder)

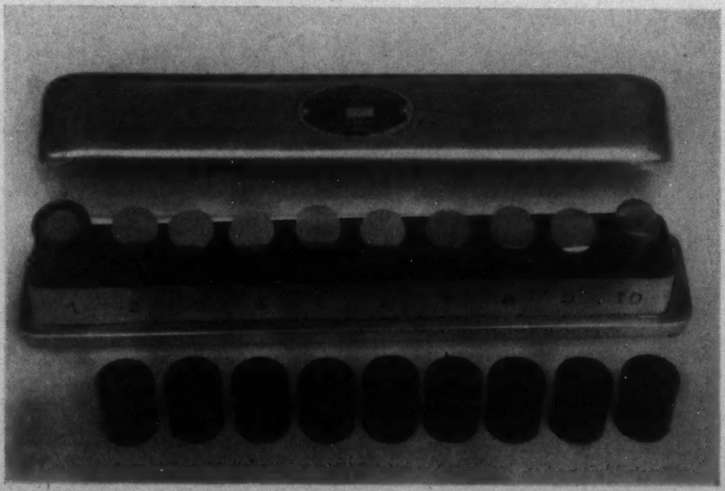

Fig. 10. A Set of Standard Martensitic Grain Fractures Developed by B. F. Shepherd. Note the increasing fineness of the fracture with increasing fracture number. (Courtesy of B. F. Shepherd)

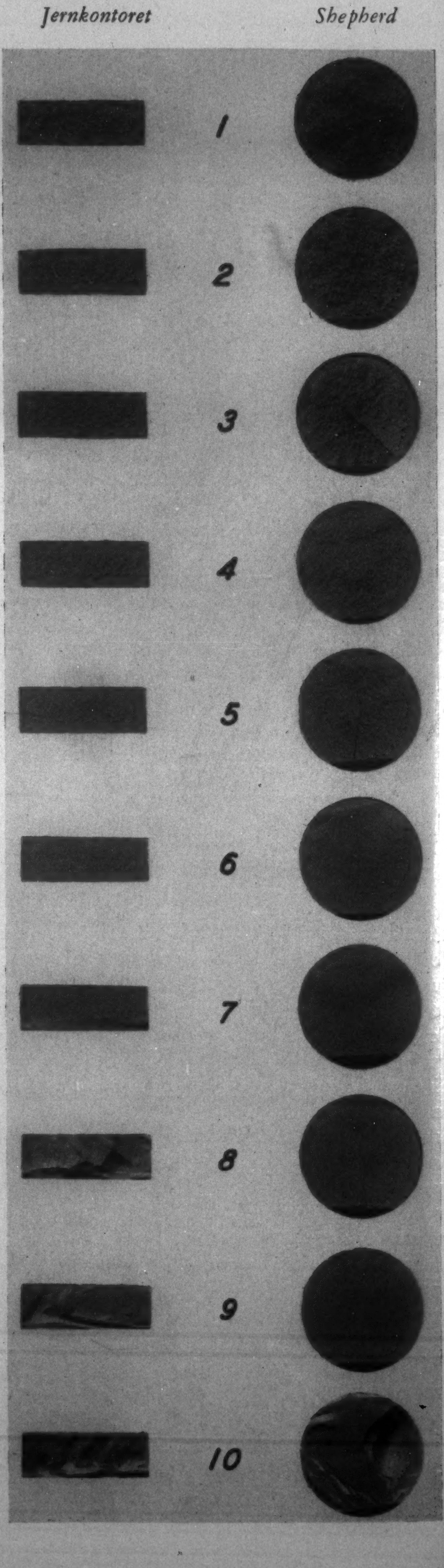

Fig. 11. Comparison of the Jernkontoret Grain Size Fracture and the Shepherd Grain Size Fracture. It is readily seen that the two scales are practically identical. (Courtesy of J. R. Vilella)

and a magnification M, the number of grains n per sq. in. at 100 diameters is

$$n = \frac{n'}{A}\left(\frac{M}{100}\right)^2. \qquad (1)$$

In order to avoid the tedious work of counting, several grain size charts for austenitic grain size were developed by steelmakers and users and from these developed the one later adopted by the A.S.T.M.[28] and now most generally used. The original A.S.T.M. chart contained eight photomicrographs at 100 diameters magnification of representative steels having grain sizes from 1 grain per sq. in. (No. 1 grain size), to 128 grains per sq. in. (No. 8 grain size). For general purposes, this chart covers the normal range of austenitic grain sizes and is sufficiently accurate for most practical purposes. Rapid estimation of the grain size number is possible by direct comparison between the chart and an image of the etched specimen on a ground glass screen at 100 diameters magnification. The A.S.T.M. grain size chart is reproduced in Fig. 8.

Davenport and Bain[35] have shown that the relationship between the number of grains n per sq. in. at 100 diameters magnification, and the grain size number is

$$n = 2^{(N-1)}, \qquad (2)$$

where N is the A.S.T.M. grain size number. Using this relationship, it was possible to extend the range of grain size numbers to values above and below the original range. Frequently it is advisable to study the grain size at magnifications other than 100 diameters. An application of equations (1) and (2) will demonstrate that for

25 diameters subtract 4 from the value given by the matching chart.
50 diameters subtract 2 from the value given by the matching chart.
200 diameters add 2 to the value given by the matching chart.
400 diameters add 4 to the value given by the matching chart.
800 diameters add 6 to the value given by the matching chart.

For some investigations it is not the grain size itself that is of significance, but more precisely the average area of the grain boundary. This arises from the fact that the transformation from austenite to ferrite and carbide or to martensite is initiated at the austenitic grain boundaries. Because of the irregularity in the size of the austenitic grains actually observed on a polished and etched specimen, it is impossible to derive a very accurate equation to yield the average grain surface area from the number of grains per sq. in. A sufficiently useful approximation, however, is possible. According to the nomenclature previously used, $1/n$ is the average area of the average grain at 100 diameters. Since the area is some constant times the square of the linear dimensions,

$\sqrt{1/n}$ = the side of a grain if it is seen as a square,
$\sqrt{1/n} / \sqrt{n}$ = the diameter of a grain if it is seen as a circle, or, in general,
$k_1\sqrt{1/n}$ = a linear dimension of the area seen, no matter what its shape may be, where
k_1 = a constant dependent upon the shape of the grain, the linear dimension selected, and the magnification.

The actual surface area of the grain is, therefore,
$k_2k_1^2\ (1/n)$ = average surface area of the grain.
By the same token, the actual volume of the grain becomes
$k_3k_1^3\ (1/n)^{3/2}$ = average volume of the grains.
Therefore the grain surface area per unit volume is

$$S = \frac{k_2\sqrt{n}}{k_3k_1} = K\sqrt{n}.$$

The actual surface area of the grain boundary per unit volume is proportional to the square root of the number of grains per unit area.

It should be noted that according to the above derivation, the boundary was counted twice, once for each of two adjacent grains. This undoubtedly leads to a satisfactory result for the present application, since the transformation zone of austenite proceeds from the same boundary into both of the adjacent grains. McBride, Herty, and Mehl[74] have shown that as the reaction proceeds, the actual grain boundary areas per unit volume become smaller. This factor, however, is of minor significance. Although it is possible to evaluate K, as has been done, by making certain simplifying assumptions as to the shape of the grains,[43, 74] nothing is gained thereby.

With the above information, it is possible to assemble the information given in Table I. A more complete table has been given by Rutherford, Aborn, and Bain[91]. Additional information on the estimation of grain surfaces and volumes has been assembled by Kaiser[97]. Snyder and Graff[99] have employed an intercept method which has the advantage of rapidity.

Table I.—Austenitic Grain Size Measurements.

A.S.T.M. Grain size (N)	Grains/sq. in. 100x (n)	Grain Surface Area/ Unit Volume (S) / (K)
—3	0.063	0.25
—2	0.125	0.35
—1	0.25	0.50
0	0.5	0.70
1	1	1.00
2	2	1.41
3	4	2.00
4	8	2.83
5	16	4.00
6	32	5.66
7	64	8.00
8	128	11.3
9	256	16.0
10	512	22.6
11	1024	32.0
12	2048	45.3

The occasional occurrence of a small grain among larger grains may be attributed to the possibility that this grain was truncated by the polished plane surface at a point where its crossectional area is small[91]. No significance, therefore, should be attributed to such an observation. Very frequently, however, true differences in grain size are noted in one specimen. (See Figs. 8 and 9).

In such instances Davenport and Bain[16] have reported both A.S.T.M. grain size numbers present and underlined the predominant one, or, more satisfactorily, reported the approximate percentages of each. On the other hand, Grossman[32] has reported only the percentage coarsened, while other investigators merely reported the average number of grains per sq. in. Each of the above methods has distinct advantages for a given type of investigation. If the point of interest centers about the coarsening temperature[87] (the temperature at which the austenite grains begin to coarsen), Grossman's method[32] appears to be most suitable. If, however, a study of the rate of decomposition of austenite is proposed, the method using the average number of grains per sq. in.[74] is better adapted. The method of Davenport and Bain[16], obviously, is an average of the other methods, and should be employed accordingly.

Recently Berglund, Hultgren, and Phragmen[79a] proposed a new grain size classification for the Swedish Ironmasters. This is referred to as the J.K.M. classification. In contrast to the A.S.T.M. classification, the J.K.M. method is based on the average area of the grains at 100 diameters magnification. The relationship between the area of the grains and the J.K.M. grain size number is

J.K.M. grain size	Grain area at 100X μ^2
0	1
1	2
2	4
3	8
4	16
etc.	etc.

according to the geometric relationship

$$A = 2^n$$

if A is the area of the grains in μ^2 at 100 diameters magnification and n is the J.K.M. grain size. Large J.K.M. numbers, therefore, refer to large grains, a system of terminology contrary to that employed by the A.S.T.M. A standard set of representative photomicrographs are employed for rapid direct comparison. Where greater precision is demanded, it was recommended that intermediate grain size numbers be reported. These are termed the leading grain size numbers and are obtained by matching the structure observed on the screen with the nearest representative photomicrograph of the standard set by varying the magnification. Accordingly, the leading grain size is calculated by

$$x = n + 2\ (\log 100 - \log f)/\log 2$$

where x is the leading grain size, n the grain size number of the matching chart, and f the magnification.

The Fracture Test

Although the actual measurement of the austenitic grain size for purposes of evaluating some factors of "quality" in steel has been used rather broadly, another method has been developed which deserves more attention than it has received. This is the fracture method developed by Shepherd [20, 21, 53] in this country and Arpi [15, 55] in Sweden. It has had a broader application in Europe than it has in America, and is commonly employed in most of the German investigations. It is more rapid than a grain size determination, and may be carried out in any plant or shop without the aid of special metallurgical equipment. Furthermore, this method allows a very fine distinction between the various types of steel. It is not impossible for a trained man to distinguish differences equivalent to 1/4 of a grain size number on the A.S.T.M. chart. Although the Swedish test, as given by the Swedish Iron Masters[55], is slightly different from that employed by Shepherd, the essentials are practically identical. For this reason, only the Shepherd test need be described.

When quenched martensitic steels are broken by impact, various types of fracture are observed[19]. In certain cases the fractured surface has a fine dull gray silky appearance. In other cases it may have a coarse fracture exhibiting numerous pinnacles and valleys indicative of fracture along the grain boundaries of coarse grains. It is also important to note that those steels that give a fine fracture when quenched from temperatures only slightly above the upper critical temperature yield a coarse type of fracture when quenched from some higher temperature. In addition, the two types of fracture are an index to the response of the steels to heat treatment. Steels that exhibit one type of fracture consistently, develop certain properties after heat treatment while the steels that exhibit the other type of fracture consistently develop other properties. These observations led to the fracture test for cataloging steels as to their response to heat treatment and other factors expressed in the term "quality." In order to evaluate small differences in "quality", a set of ten martensitic fractures ranging uniformly from the coarsest to the finest were assembled by Shepherd. These fractures were numbered consecutively from No. 1, the coarsest, to No. 10, the finest. It was then possible to assign a "quality" number, a fracture number, to any steel by direct visual comparison of the fracture of a steel with the corresponding fracture in the standard set. A typical standard set is shown in Fig. 10.

The following recommendations for the test are a condensation of those given by Shepherd[53].

1. Preparation of Test Pieces.
    a. Obtain a representative sample of steel.
    b. Machine 4 specimens 3/4 in. ± .001 in. in diameter and 3 in. long.
    c. Stamp the specimens 1400, 1500, 1550, and 1600 respectively.
2. Hardening.
    a. An electric furnace should be used for heating and should closely reproduce the rate of heating in practice. Each sample is heated to the temperature in deg. F. stamped on the piece.
    b. The samples shall be quenched in a standard verti-

cal jig (2 in. pipe overflow) flushed with a 10 per cent brine solution at room temperature. Care must be taken to eliminate the effect of the tongs as much as possible.

3. Fracture Grain Size.
   a. Specimens are notched midway between the ends to a depth of not over 1/16 in. and broken.
   b. The grain size is determined by comparing and matching the case or outer zone of one of the broken halves with the fracture standards.

The results of such a test yield 4 numbers corresponding to the 4 temperatures of the test. If the temperature of the heat treatment to be used in practice is known, only one test, namely that which agrees in all details with the actual heat treatment, need be made.

Another feature of the test is a measure of the penetration of hardness, but it appears suitable, in the light of present knowledge of the subject, to include this under the category of properties that are influenced by the austenitic grain size. Therefore this will be referred to later, under hardenability.

The Swedish method is practically identical with Shepherd's method (See Fig. 11). Furthermore, Vilella and Bain[79] have shown that the two standards of fracture are identical for all practical purposes. If comparison with the Swedish standards yields a No. 6 fracture on the Swedish Iron Masters' scale, it has a No. 6 fracture when compared with Shepherd's standard fractures. This is very important for an unambiguous interchange of ideas between European and American metallurgists. In addition to this fortunate identity in standards, Vilella and Bain demonstrated, by special martensitic etching methods, that the martensitic fractures occur essentially along the boundaries of the former austenitic grains (See Fig. 12). This accounts for the relationship between "quality" and fracture, and is proof that the fracture method is only another way of measuring the austenitic grain size. It must be admitted, therefore, that the martensitic fracture is as fundamental as is austenitic grain size as an index of "quality."

A comparison of the A.S.T.M. austenitic grain size number and Shepherd's standard fracture number by Vilella and Bain[79] has shown that there is very close agreement between the two. The deviation is generally less than one-half a grain size number (see Fig. 13). This agreement [66] is very fortunate, as a No. 6 grain size on the basis of one of the above three standards is, for all practical purposes, a No. 6 on the basis of the remaining standards. No confusion need result, therefore, if the standard is not mentioned.

The human eye appears to be more sensitive to small variations in fracture than to small variations of grain size when observed as an image on a screen. For this reason, a higher precision is possible with the fracture method. Nevertheless, if the austenitic grains exhibit duplexed grain sizes, large and small grains simultaneously, the entire fracture may appear coarse to the uninitiated observer. With practice, however, it is possible to distinguish between a truly coarse grained steel and one which has duplexed grain sizes. In cases of doubt, it is advisable to determine the austenitic grain size by comparison with the A.S.T.M. chart.

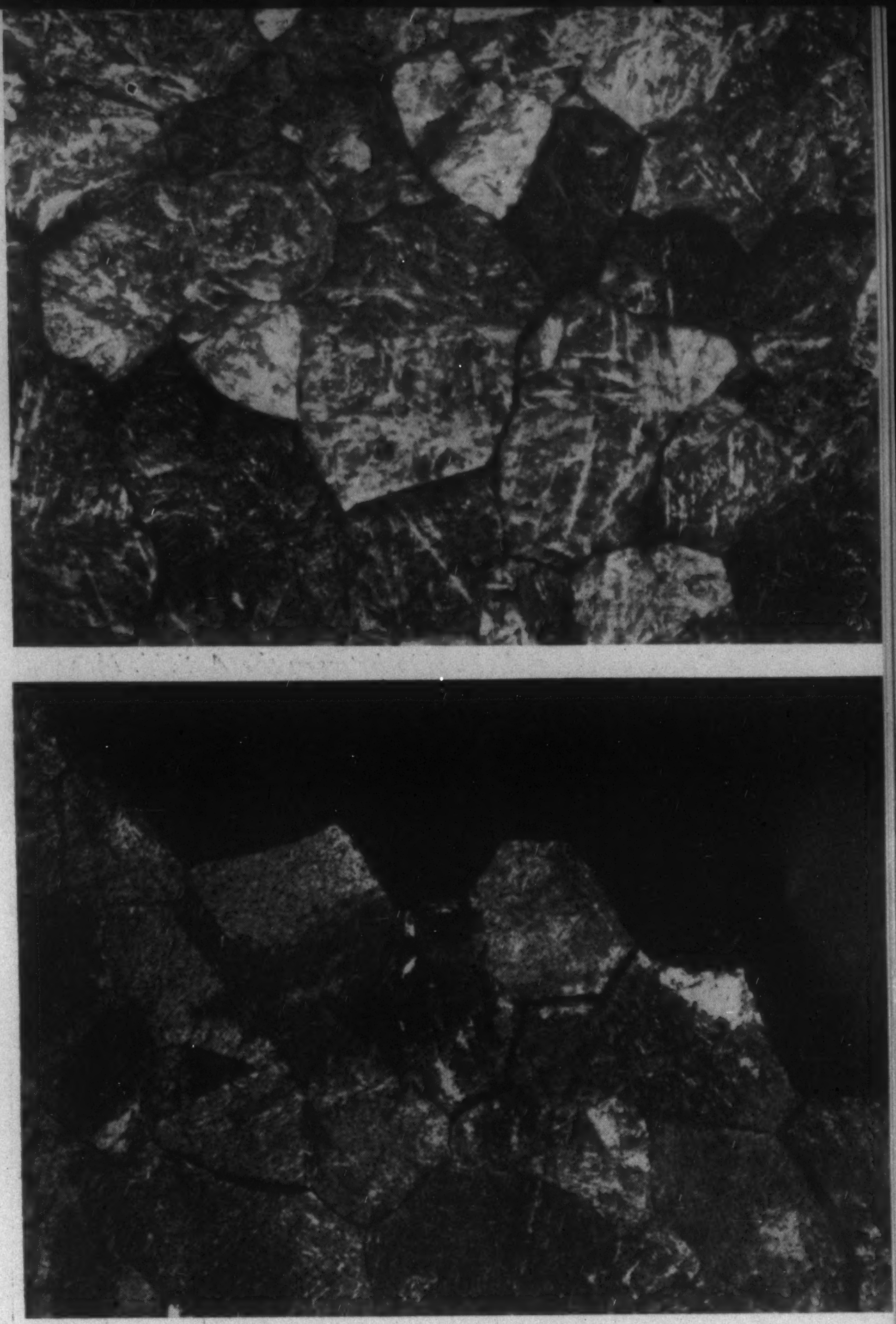

Fig. 12. Photomicrographs of Quenched Steels Etched with a Special Martensitic Etchant to Reveal the Austenitic Grain Size. Note that the martensitic fracture occurs along the boundaries of the former austenitic grains. (Courtesy of J. R. Vilella)

Austenitic Grain Size and Rates of Transformation

Undoubtedly the most important effect of austenitic grain size on the "quality" of steel is its influence on the rate of transformation of austenite to ferrite and carbide or to martensite. The important indices of "quality," such as hardenability and toughness, are essentially the result of this influence on the rate of transformation.

The present detailed knowledge of rates of decomposition of austenite is a result of the classic investigations of Bain and Davenport and their associates[16, 22, 23, 35]. Although there had been numerous previous

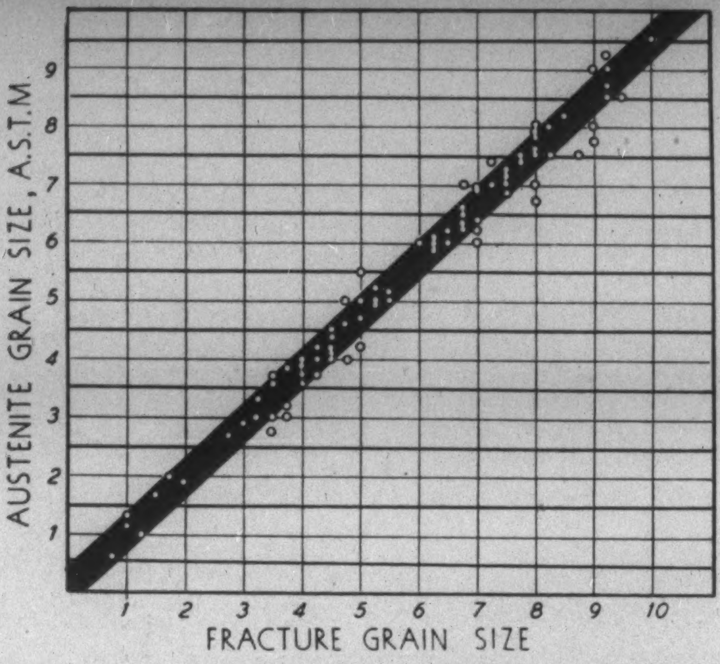

Fig. 13. Relationship between the A. S. T. M. Grain Size and the Fracture Grain Size. (Courtesy of J. R. Vilella)

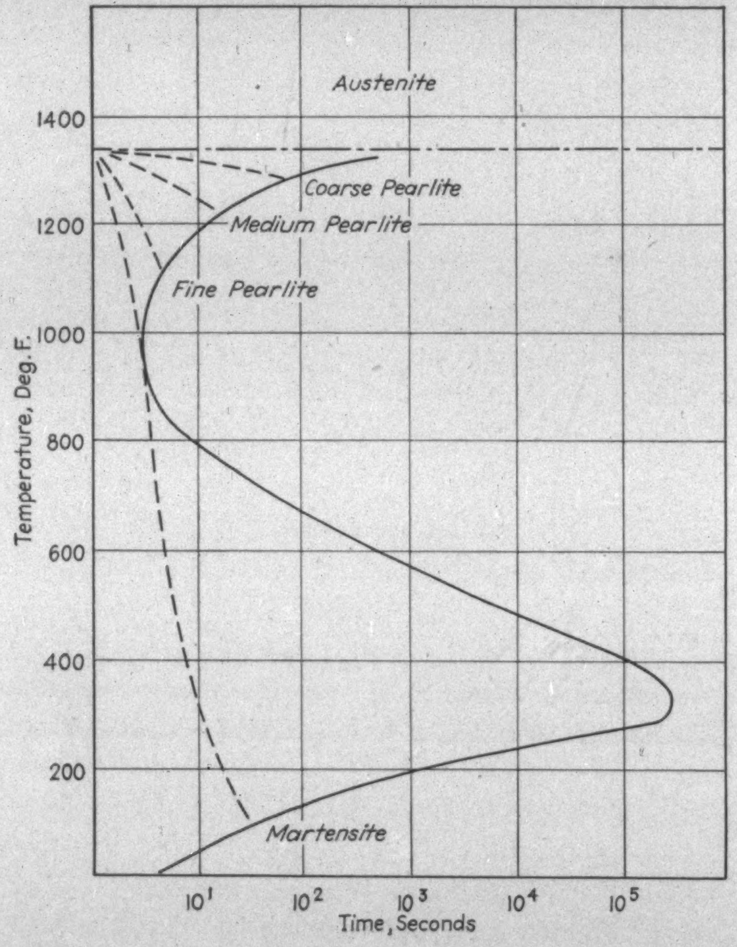

Fig. 14. Schematic Representation of the "S Curve" of Bain and Davenport. The S curve indicates the terminus of the isothermal induction period. Decomposition of austenite commences when cooling or quenching curves, shown by broken lines, intersect the S curve. The microconstituents formed, however, depend upon the temperature at the intersection of the two curves as indicated above. The critical quenching rate curve is tangent to the S curve at the point of minimum induction period, about 1000 deg. F. All rates exceeding the critical quenching rate yield martensite. The above curve does not show the elapsed time from beginning to end of transformation at the various temperatures.

investigations on the transformation of austenite, Bain and Davenport [35] employed a simple method that could readily yield quantitative results. Small samples of steel were heated above the critical temperature until they had transformed entirely into austenite. Then they were rapidly cooled to some subcritical temperature, usually by immersion in a lead or salt bath, and held at this temperature for various periods of time. This treatment was followed by rapid quenching in brine at the specified time intervals. It was possible to determine, by microscopic examination, the rate of the transformation.

They found that for any given constant temperature, the decomposition of austenite did not start immediately. (The interpretation that nothing happens during this time is probably incorrect. The steel is overcoming its inertia to transformation. This is a common feature of all chemical reactions, especially chain reactions, and the time which elapses is called the induction period of the reaction.) After a certain interval of time the induction period terminated and the first evidences of ferrite or carbide could be seen. The induction period for a given steel was found to be dependent upon the temperature at which the reaction was studied. For temperatures near the eutectoid temperature the induction period extended over as much as a 5-min. interval for some steels. For successively lower temperatures the induction period diminished until it reached a minimum at approximately 1000 deg. F. At this temperature the induction period became an interval of only a few seconds. As the temperature of the bath was lowered the induction period increased and reached a maximum which was of the order of magnitude of 1 hr. for some steels. At still lower temperatures the induction period again diminished, and the structures formed were found to be martensite or its decomposition products. The time-temperature relationship for the transformation is indicated graphically in Fig. 14.

The induction period has a very special significance in hardening steel. In order to obtain a hard martensite structure it is necessary to quench sufficiently rapidly to pass the region at about 1000 deg. F., where the induction period is short, before transformation to ferrite and carbide begins. If some transformation occurs in this range the steel will have soft spots that will have a hardness of fine pearlite which is about C-28 Rockwell. If the steel can be quenched

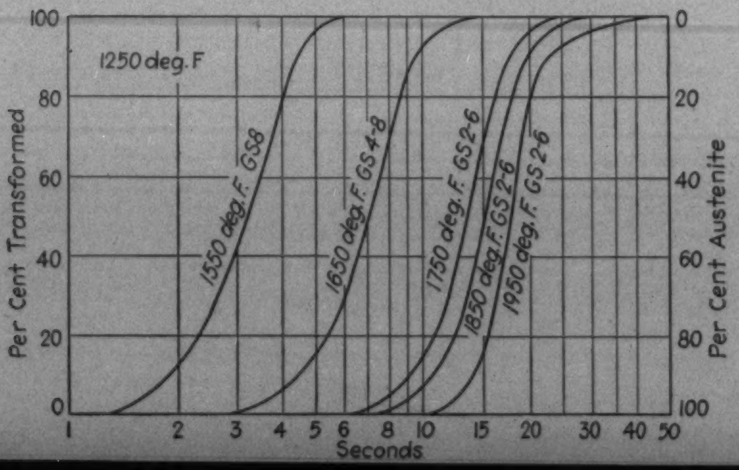

Fig. 15. Left: Curves Illustrating the Influence of Austenitic Grain Size on the Induction Period and the Rate of Decomposition of Austenite at 1250 deg. F. Samples of the same heat were treated at various temperatures to develop various austenitic grain sizes. Each specimen was then quenched in a lead bath at 1250 deg. F. and the per cent austenite transformed at given time intervals was determined. Note that as the grain size increased from 8 to 2-6 the induction period also increased. In addition, it may be seen that the time required for the transformation itself is greater for the coarser grained steels. (Courtesy of E. S. Devenport)

to temperatures below this range before any transformation has started, it will, under average conditions, transform to martensite because of the longer induction periods at the lower temperatures. This quenching rate is synonymous with the term "the critical cooling rate."

After the induction period has elapsed, transformation begins to take place. Bain [23] found that, in general, the rate of this reaction is closely analogous to the rate of a unimolecular chemical reaction. This means that at any time the rate of transformation is proportional to the amount of untransformed austenite at that time. Expressing this in the form of an equation yields

$$-dx/dt = Cx$$

Where x = fraction of untransformed austenite,
$-dx/dt$ = the rate at which austenite is disappearing and
C = the proportionality constant.

Integration of this equation yields

$$\text{Log } x = -Ct,$$

if the time, t, is measured from the end of the induction period. The correlation between the theoretical curve of the above equation relating the fraction of untransformed austenite to the time agrees extremely well with the actual experimental results over the greater part of the reaction. The theoretical rate curve for a unimolecular reaction does deviate somewhat from the actual rate curve at the start of the reaction. This may be due, in part, to the changes that must occur during the induction period. In spite of this deviation, it is advisable to describe the reaction in terms of a unimolecular reaction, remembering, of course, that the analogy is not quite perfect. Hence, the constant C becomes the index for the transformation.

C is a function of the temperature. In addition, C depends upon the concentration and kind of alloying elements. Additions of manganese to the steel retard the rate of transformation or, in other words, decrease the value of C.

Bain [22] found that changes in the austenitic grain size of a given steel influenced pronouncedly (1) the duration of the induction period and (2) the rate of reaction or the value of C. This is readily understandable on the basis of the information given, concerning the methods of revealing the austenitic grain boundaries. The boundary regions are more reactive than the core of the grain, and consequently have less inhibition to transform to ferrite and carbide under suitable conditions than does the grain itself. For a given volume of steel, then, the inhibition to transform becomes less as the grain boundary areas increase. Fine austenitic grained steels, therefore, have shorter induction periods than coarser austenitic grained steels. Furthermore, the austenitic grain boundaries serve to promote nucleation for the transformation. Consequently, finer austenitic grained steels will transform over a greater fraction of the volume than the coarser austenitic grained steels. The rate of transformation (the reaction constant, C) of fine austenitic grained steels is, therefore, greater than for the coarser austenitic grained steels. The influence of austenitic grain size on the induction period and rate of transformation at 1200 deg. F. is shown in Fig. 15.

Herty, Lightner and McBride [43] and McBride, Herty, and Mehl [74] have employed the above principles in their study of the rate of formation of ferrite. In these investigations, various values of C, the reaction constant, were obtained for various values of n, the number of grains per sq. in. at 100 diameters magnification, for a single steel at the same temperature. When the reaction constant, C, was divided by S, the surface area of the grains per unit volume, a constant was obtained which was essentially independent of the grain size. The obvious conclusion is that the rate of reaction is practically proportional to the surface areas of the grains per unit volume. The authors of this investigation extended their work and showed that the ratio of the amount of ferrite formed by normalizing to the amount formed by annealing the same steel, F_n/F_a, is a linear function of S, the surface area of the grains per unit volume. In this way it was found that contrast between the annealed and normalized structures was the greatest for the coarser grained steels. Furthermore no difference between the normalized and annealed structures was noted after the value of S, or the number of grains per square inch n, exceeded a definite value which depended upon the size of the specimen. This is illustrated in the following table:

Cube edge	¼ in.	½ in.	1 in.	2 in.
n	715	179	41.5	11.2

According to this table, there will be no difference in the normalized and annealed structures for a two inch cube, if the number of grains per sq. in. at 100 diameters exceeds 11.2 for the steel investigated.

The information correlated in this section shows that the austenitic grain boundary area per unit volume influences the rates of transformation and the induction period. Finer grained steels transform sooner and more quickly than coarser grained steels. Therefore such steels must be quenched more rapidly than the coarser grained steels in order to obtain full hardening. This reveals the important differences between fine and coarse grained steels. Many of the differences in "quality" such as hardenability and toughness, etc., as will be shown later, are directly attributable to these differences in rates of reaction. Studies on the effect of the austenitic grain size on the rate of transformation of cast irons have yielded similar results [75, 89, 98].

(To be continued)

All references published in March issue.

Electron Diffraction Examination of Metals

CONCLUDED FROM MARCH ISSUE

Operation

After the specimen holder is placed in position and the plate holder loaded, the camera is evacuated until the thermocouple vacuum gage indicates that a satisfactory degree of vacuum has been attained. The filament circuit is closed, the voltage applied to the slit-plate in front of the filament and the filament rheostats manipulated until an emission current of 100-150 microamperes is indicated. With the room darkened, the slits are adjusted until the direct electron beam has its maximum intensity as viewed on the fluorescent screen. During this operation the specimen carrier is raised so that the beam passes uninterruptedly below the specimens. (For the initial adjustment of the camera after replacing a filament, each slit holder is provided with a 3/16-sq. in. opening on the same holder as the narrow slit is mounted. With these large openings in position, the filament assembly is aligned so that the electron beam, projected directly from the slit-plate in front of the filament, strikes the center of the fluorescent screen with maximum intensity. With this adjustment made, the several narrow slits are moved successively into the path of the beam so as to yield a narrow beam of the maximum intensity.)

by M. L. FULLER

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After the slits are adjusted, the specimen carrier is lowered so that one of the specimens partly intercepts the beam. The angle that the specimen surface presents to the beam is first adjusted so that the surface is parallel to the oncoming beam and then a sufficient rotation of the specimen carrier is effected to make the angle of grazing incidence about one-half degree. These manipulations of the specimen can be governed very nicely by observation of the fluorescent screen. With the angle of the specimen adjusted, the carrier is raised or lowered vertically until the maximum illumination of the fluorescent screen is obtained in that section of the screen where the diffraction pattern should fall. In dealing with polycrystalline material, one can observe but rarely the diffraction rings themselves. Having completed these adjustments, the pattern may be recorded on the photographic plate.

The electron beam is cut off by opening the filament emission circuit, the plate holder moved into position, and the filament emission circuit closed for the duration of the exposure.

Exposure times vary from 5 to 30 sec., depending upon circumstances. The actual electron current passing from the slit plate in front of the filament to the anode is not known but it is much less than the total filament emission of 100 to 150 microamperes. The impressed voltage is usually of the order of 40 kv. and its actual value is determined for each plate by photographing thereon the pattern of some standard substance, pure zinc oxide being used in most cases.

By the use of a rectangular diaphragm at the entrance to the plate holder chamber, a series of exposures of the several specimens mounted on the specimen carrier can be recorded on each plate. Each specimen must be properly positioned by the manipulations and fluorescent screen observations described above, before each photographic exposure. During these adjustments the plate holder must be moved back into its light-tight chamber in the position shown in Fig. 2.

In order to record upon the plate an accurately measurable record of the position at which the undeviated beam strikes the plate when taking reflection type photograms, the following procedure is employed. After each exposure, the specimen carrier is raised so as to permit the uninterrupted passage of the beam and simultaneously the plate holder is moved along its guides within the plate holder chamber. After thus producing a base line on the plate, the filament emission circuit is broken. Such base lines are shown on several of the reflection type photograms reproduced in this paper.

Preparation of Specimens

The principal requirement of a transmission type specimen is that the film of specimen be thin enough to transmit the electron beam, a requirement that imposes a maximum thickness of about 500 A. U. (0.00005 mm.). For the technique of preparing such films, the reader is referred to the literature on electron diffraction, much of which reports work done by the transmission method.

Since most industrial applications of the electron diffraction technique will necessarily deal with the examination of the surfaces of massive specimens, the writer will devote more attention to the preparation of such specimens. It was pointed out by G. P. Thomson[4] that the surface "reflection" of electrons takes place by the transmission of electrons through minute projections at the surface of the specimen. The thickness of the section of such a projection that is presented to the electron beam must be 500 A. U. or less in order to transmit the electrons. Thus, a specimen that is too smooth or one too rough is unsuited for electron diffraction. An intermediate and optimum surface texture should be produced if possible. This may be produced by abrading the surface with emery paper, a wire brush or a file, the particular conditions producing the best results being determined by trial. In the examination of specimens showing corrosion films, such films will, in general, be of a suitable texture without special treatment although in investigating the corrosion of a metal it may be desirable to prepare the original clean metal surface by some procedure known to produce a surface texture suitable for electron diffraction.

Powdered samples may be prepared by pressing the powder against a metal block, initially roughened with coarse emery paper. The surface of the powder should be made smooth and reasonably flat, an operation which can be carried out by stroking with a razor blade.

Examples of Photograms— Transmission Photograms

When the specimen can be obtained in a form thin enough to transmit the electron beam, complete ring patterns of the Debye-Scherrer type are obtained. Fig. 5 shows a transmission photogram of zinc oxide (A) and of gold leaf (B). The zinc oxide specimen was prepared by allowing the fume from a zinc oxide furnace to deposit on an 80-mesh nickel screen. The zinc oxide particles deposit upon one another, bridging the openings in the wire mesh. Discrete diffraction spots from individual zinc oxide crystals are visible on the diffraction rings. An enlarged shadow of the wire mesh may also be seen on this photogram. The gold leaf specimen was of commercial gold leaf that had been thinned to about half its original thickness by floating upon a 2 per cent potassium cyanide solution.

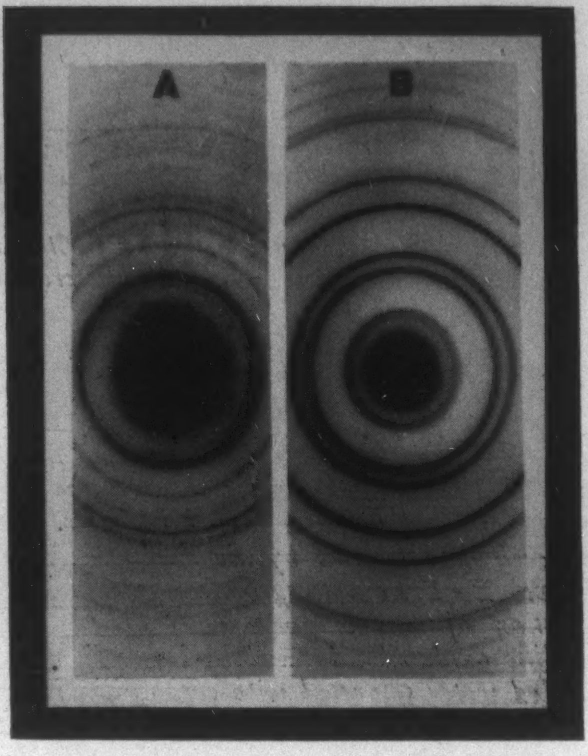


Fig. 5. Transmission Electron Diffraction Photograms. A—Zinc Oxide; B—Gold. Note: *The relative intensities of the diffraction maxima in the electron diffraction photograms reproduced in this paper are not exactly the same as in the original negatives owing to the necessity of controlling the illumination in printing so as to yield clear reproductions.*

The Debye-Scherrer type patterns obtained from polycrystalline specimens are analogous to the corresponding X-ray patterns. The interplanar spacings of the crystal lattice corresponding to each diffraction ring may be calculated from the ring diameters, the specimen to photographic plate distance, and the electron wave length, according to the Bragg law of X-ray diffraction.

The interpretation of electron patterns in order to identify unknown substances is carried out by comparison of the unknown pattern with the electron (or X-ray) diffraction patterns of known substances. It is not intended to discuss the matter of calculations and interpretation further in this paper since such information is available in the literature.

Thin Films on Metal Surfaces

Reflection type photograms taken with the purpose of detecting films on metal surfaces are illustrated in Fig. 6. In the preparation of zinc and zinc alloy surfaces for electroplating the surface is subjected to a series of cleaning treatments. The first step is to buff or polish the surface. Since this leaves a thin grease film the object is usually degreased, following the buffing, with a solvent. Photogram A of Fig. 6 is from a zinc alloy die casting after degreasing with trichlorethylene. The electron pattern shows that the

Fig. 6. Thin Films on Metal Surfaces. A—Film resulting from action of trichlorethylene degreasing solvent on zinc alloy die casting. B—Brittle copper-zinc alloy found on inner side of blistered plating.

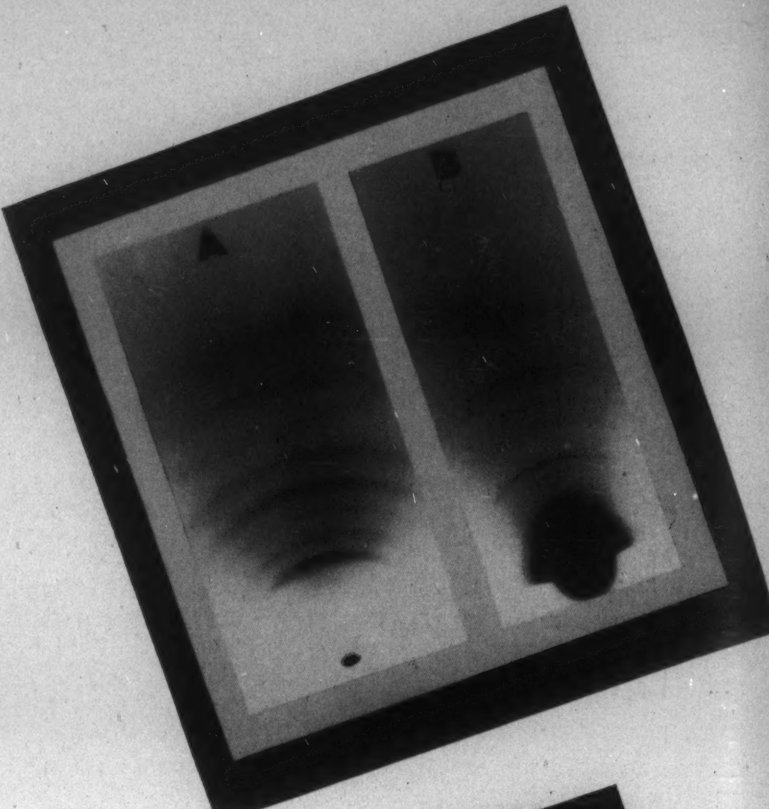
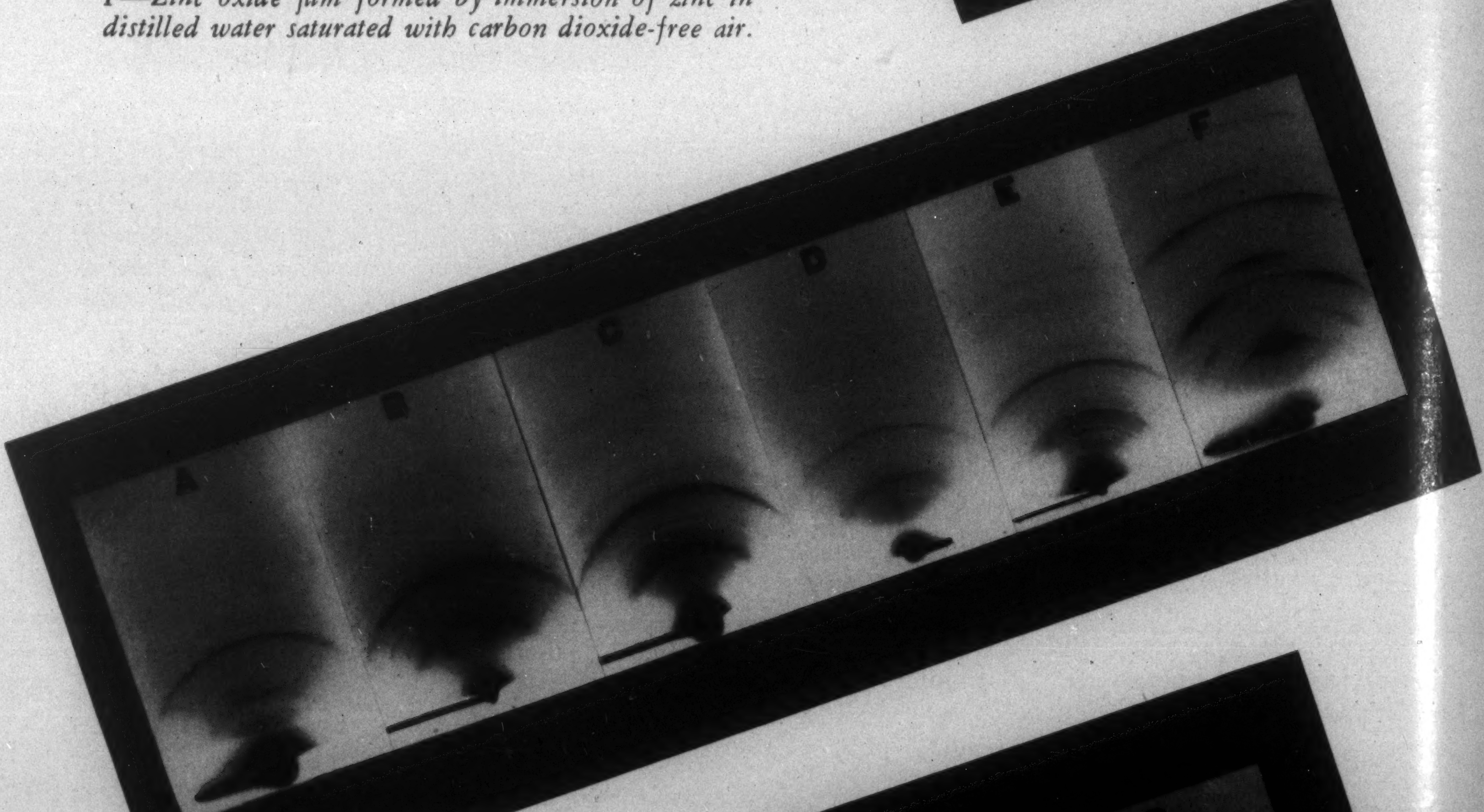


Fig. 7. Corrosion Films on Zinc. A, B and C—Basic zinc carbonate formed by the immersion of zinc in distilled water saturated with air. D—Basic zinc carbonate formed by the atmospheric corrosion of zinc. E—Basic zinc carbonate formed by the precipitation of zinc sulphate with sodium carbonate. F—Zinc oxide film formed by immersion of zinc in distilled water saturated with carbon dioxide-free air.



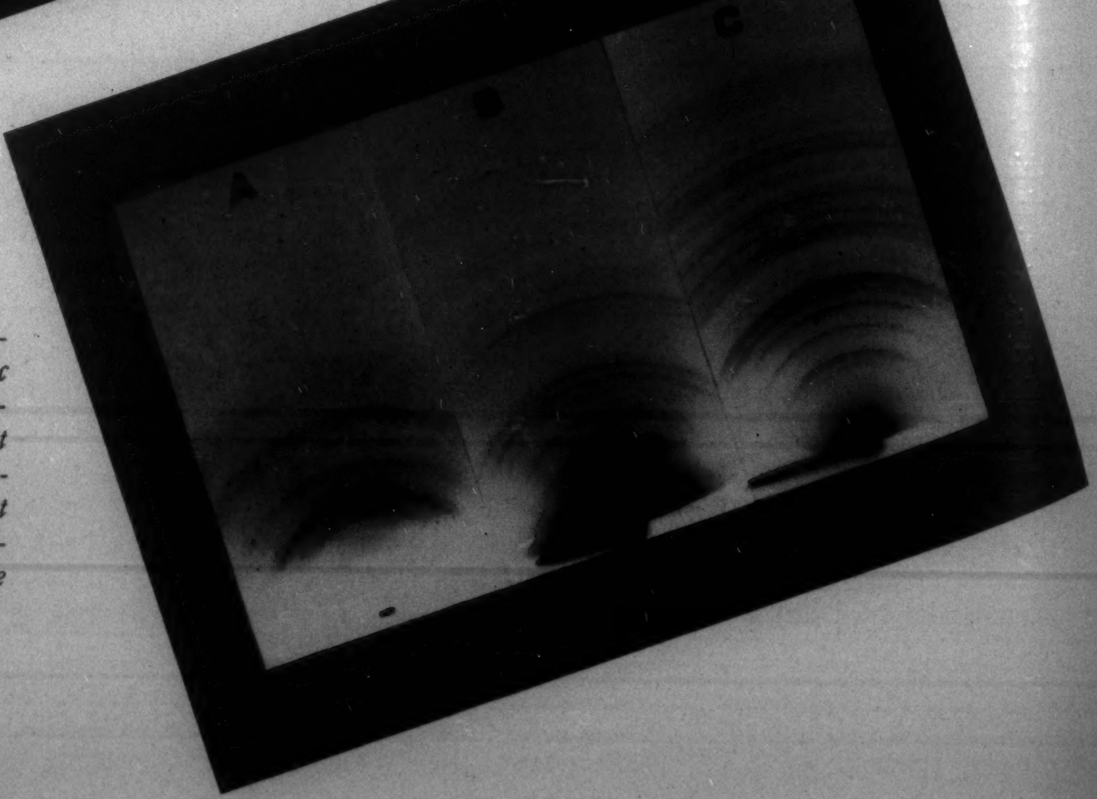


Fig. 8. X-Ray and Electron Diffraction Photograms of Zinc Dust Pigment. A—X-ray diffraction photogram of zinc dust pigment. B—Electron diffraction photogram of zinc dust pigment. C—Electron diffraction photogram of zinc oxide pigment.

grease has been removed leaving the surface coated with a reaction product of the solvent upon the zinc. This film is not always formed by trichlorethylene degreasing and when it is formed it is normally removed in the next step in the cleaning process, usually electrolytic alkaline cleaning.

The exact nature of this film has not been determined but investigation has disclosed that it is specific for the action of the solvent on zinc.

Photogram B of Fig. 6 was obtained from the inner surface of a plated coating that had blistered away from a zinc alloy casting. The plating was of the copper-nickel-chromium type, the copper being deposited first. The electron pattern indicates the inner surface of the blister to be a brittle copper-zinc alloy. Evidently the electrodeposited copper had diffused into the zinc casting, forming a brittle phase within which stratum the plating separated from the casting.

Corrosion Films on Zinc Surfaces

With the aid of the electron diffraction camera the film formed by the ordinary atmospheric corrosion of zinc or by water immersion has been identified as basic zinc carbonate. The immersion of a clean zinc surface in distilled water saturated with air results in the formation of a complete coating of the basic zinc carbonate in ½-min. immersion. The latter film is too thin to be visible to the eye but is readily detected by electron diffraction examination. The formation of a complete film by atmospheric exposure requires a longer period, the length of which depends upon the relative humidity being ordinarily of the order of several days. The immersion of zinc in water saturated with carbon dioxide free air results in the formation of a coating of zinc oxide. It is planned to publish the complete details of this corrosion investigation in a separate paper but some of the electron diffraction photograms will be reproduced here in order to illustrate the possibilities of the electron camera in this type of investigation.

Fig. 7 is a reproduction of six of the photograms taken in this study. Photograms A, B, and C are of the basic zinc carbonate formed by the immersion of zinc in distilled water saturated with air. Photogram D is of a zinc specimen corroded by atmospheric exposure. Photogram E is of chemically prepared basic zinc carbonate formed by the interaction of solutions of zinc sulfate and sodium carbonate. It will be noted that the photograms A, B, C, D, and E are very similar. The differences among the several patterns are due to differences in degree of carbonation and hydration and to differences in the physical structure of the several specimens. Photogram F shows the diffraction pattern of the zinc oxide film formed by immersion of zinc in water saturated with carbon dioxide free air. The zinc oxide crystals of this particular film are preferentially oriented with their basal planes parallel to the zinc surfaces. The zinc oxide films produced in this manner are sometimes oriented at random, however.

Zinc Oxide Coating on Zinc Dust Particles

Illustrative of the low penetrating power of the electron beam is the type of photogram given by a zinc dust pigment. This pigment is made by the condensation of zinc vapor in a manner similar to the process for manufacturing sulfur powder. Fig. 8, photogram B is an electron diffraction photogram of a specimen of zinc dust. This sample contains only 2 per cent zinc oxide but the electron pattern arises principally from the zinc oxide. Each zinc dust particle is coated with a layer of zinc oxide sufficiently thick to prevent the electron beam from being diffracted by the metallic zinc particle. Photogram C is an electron photogram of pigment zinc oxide reproduced for comparison. An X-ray diffraction pattern of the zinc dust is shown at A. The X-ray pattern of the zinc dust arises principally from the metallic zinc, the X-rays being of sufficiently high penetrating power to yield diffraction effects from the whole of each particle. (The reproduction of the X-ray pattern has been reduced to approximately the same dispersion as the electron patterns).

In this experiment the electron technique demonstrates in a very striking manner that the zinc dust particles are coated with zinc oxide. That this should be the case is not in itself surprising but the photograms are reproduced here to illustrate the applicability of electron diffraction to this type of study.

Acknowledgment

The writer expresses appreciation for assistance in the design of the apparatus and in its application to other members of the Research Division, particularly E. A. Anderson under whose direction the work was performed.

[4] G. P. Thomson. "Wave Mechanics of Free Electrons". McGraw-Hill Book Co., New York, 1930, page 158.

[5] G. I. Finch & A. G. Quarrell. *Proceedings of the Royal Society*, Vol. 141, 1933, page 398.

World Electric Steel Output

by EDWIN F. CONE

THE PRODUCTION OF STEEL in electric furnaces has expanded in recent years to relatively large proportions, both in the United States and in other countries. There is keen interest in this growth. So much so that a survey of available data covering the world output has seemed advisable and timely. This has been made possible due to the excellent and reliable statistics of world production of iron and steel which are published annually by the British Iron and Steel Federation of London, England.

It is generally admitted that the electric steel industry in the United States started on a commercial scale about 1913. This may also be accepted for most other major producing countries. Therefore the world's electric steel industry may be regarded as 25 years old with the close of 1937.

World Output Includes 12 Nations

An analysis of the British data, referred to above, has been made with the result that Table I has been compiled. According to this there are 12 countries which are today producing steel in electric furnaces —both arc and induction. These are: The United States, The United Kingdom, France, Germany (including the Saar), Belgium, Luxemburg, Japan, Italy, Poland, Sweden, Canada and India.

Undoubtedly Russia produces steel in electric furnaces, but the British data do not include this item for that country. This is also true of Norway and possibly also of one or two other countries. However, the data are fairly complete and afford a measure of the progress of the world's electric steel industry.

In Table I, the electric steel production of these countries is given for 1913, 1929, 1934, 1935, 1936 and 1937, for comparison. Since 1934 the Saar has been included in the German statistics. The data are in both gross and metric tons, 2240 and 2205 tons respectively, both being practically the same for purposes of comparison.

In 1913 the world production of electric steel was only 172,746 tons. In 1929 this had grown to 1,871,265 tons or a little over tenfold. This was the predepression high. In 1930 this total declined to 1,490,200 tons and then to 1,252,600 tons in 1931. As late as 1934 there was no recovery. In 1935, however, the world total exceeded 2,200,000 tons for the first time at 2,226,031 tons. And from then on the expansion was rapid until, in 1937, the latest data available, the same countries made over 3,250,400 tons. (Attention should be called to the fact that, in the 1937 data and in one or two other years, estimates had to be made.) Thus in 25 years the world output has increased from 172,746 tons to 3,250,499 tons, or over 175 per cent—a significant expansion.

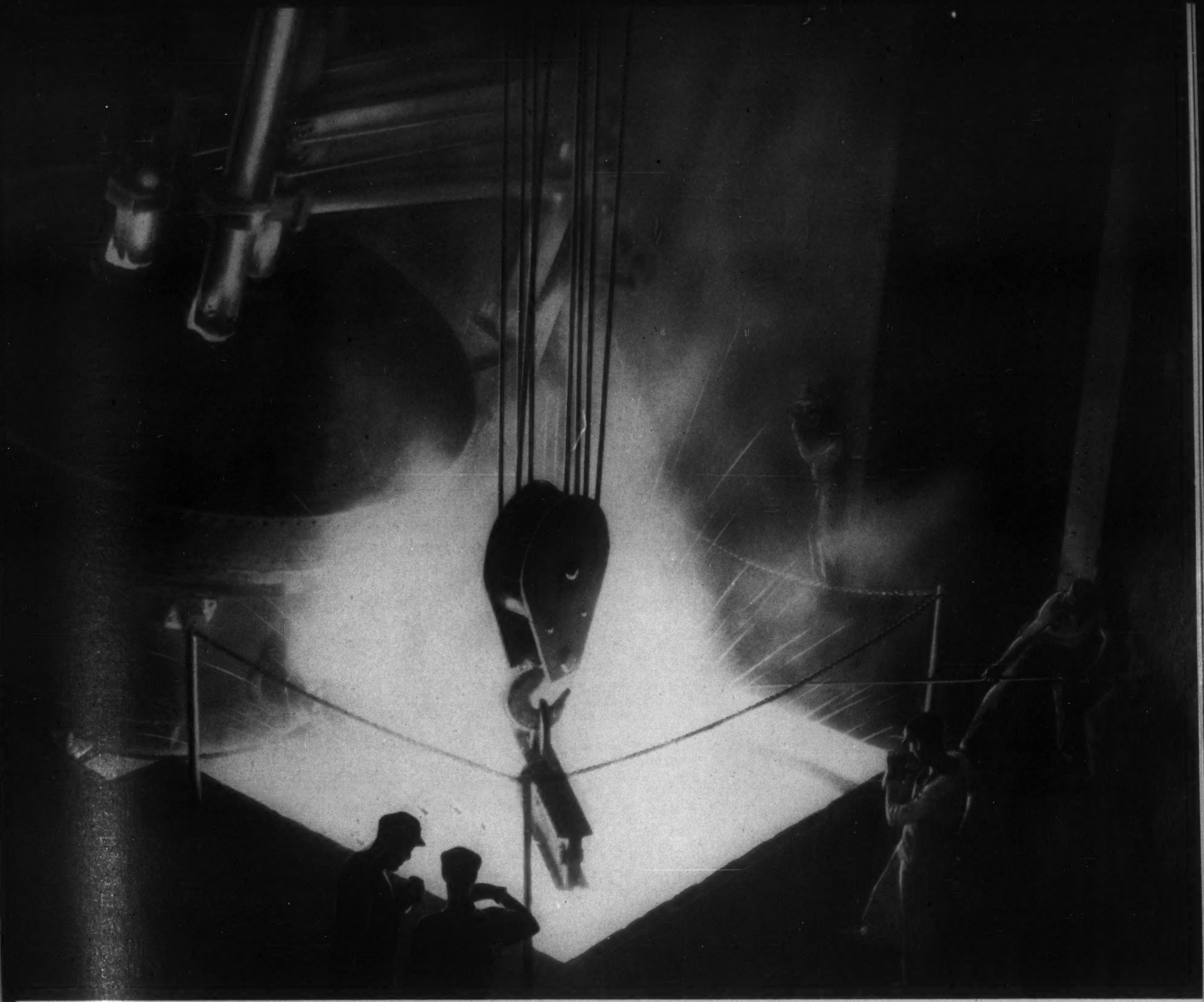
Record of Leading Countries

Reviewing the records in some of the countries individually, certain highlights appear.

United States: The high point in the electric furnace steel was in 1929. This was not equalled in 1937 though, were the actual electric steel casting production included, it is probable that the total would equal, or perhaps slightly exceed, that for 1929. The 1937 total of 845,537 tons was the largest for any of the 12 nations.

There is no published statement of the actual number of electric steel furnaces in the United States. The number is large, including both arc and induction. The latter type has gained in the number of installations in recent years.

Germany: Prior to 1924 the electric and crucible steel output of Germany were each given but since



Pouring Stainless Steel from an Arc Electric Furnace. This is one of 4 such furnaces at the plant of the Rustless Iron & Steel Corp., Baltimore. Arc furnaces include the Heroult, Lectromelt, Swindell and others.

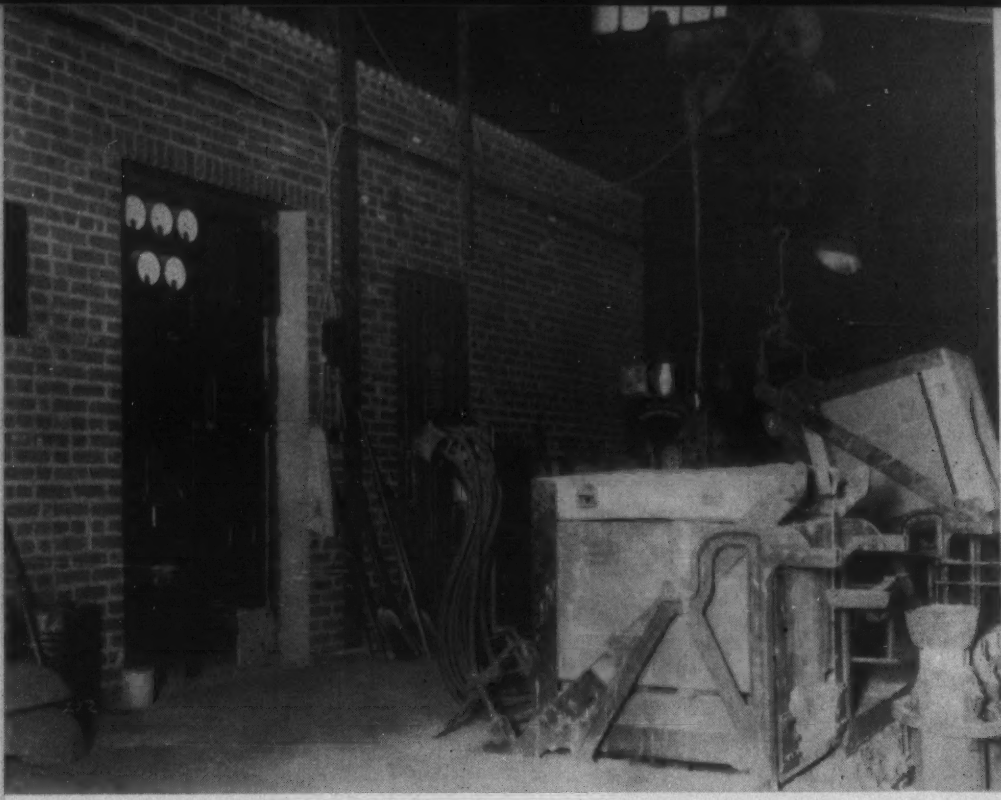
then they have been combined so that totals in Table I include crucible steel. Castings are probably not included. Since the crucible output had constantly diminished down to 16,348 tons in 1923, it is probable that this trend has continued, so that the data are largely electric steel. The 1937 output at 700,369 tons ranks second among the totals and is Germany's largest to that time—in contradistinction to the course in the United States. In 1913 Germany led the world in electric steel production at 88,881 tons.

Italy: Next in magnitude in 1937 is Italy at 610,000 tons, estimated. The total was 591,275 tons in 1936. In 1917, the first data for this country, the total was 36,948 tons, so that the expansion in 20 years to over 600,000 tons has been striking. This is natural since Italy has a wealth of electric power as against limited resources of raw materials for open-hearth and pig iron production. The data reported do not include castings.

France: The growth of the French electric steel industry has been from 21,124 tons in 1913 to 315,990 tons in 1937, a relatively large expansion, ranking fourth among the 12 countries. These data include castings. Paralleling Germany, France was one of the early pioneers in electric steel. The 1937 total was a new high for France.

United Kingdom: Expansion of the British electric steel industry has been relatively small or from 22,000 tons in 1915 to 215,400 tons in 1937 or about tenfold. It ranks sixth among the nations reviewed. The total includes castings—60,800 tons in 1937, with 154,600 tons of electric steel ingots. In 1915 castings were only 2,000 tons with 20,000 tons for ingots.

Japan: The latest data available from Japan state that the electric furnace steel production was 173,042 tons in 1935. Estimates place the 1937 total at 210,000 tons. Only ingots are included in these totals.



Two Ajax-Northrup High Frequency Electric Furnaces. Considerable high grade alloy and other steels are made in this type.

The 1937 estimate shows a rather remarkable expansion above a total of only 4,329 tons in 1918. This would place Japan seventh among the nations considered.

Sweden: The growth of the electric steel industry of Sweden has been remarkably rapid or from 2,276 tons in 1913 to 253,800 tons in 1937, with a ranking of fifth among the 12 nations. These data include castings. Like Italy, Sweden has ample electric power resources and the expansion is what might be expected.

Canada: Data for electric steel furnace production runs back to 1905 when 3,320 tons of ingots and castings were produced. In 1913 the total was only 449 tons. During the war 107,376 tons were made in 1918 but this has not been approached since,—the output has been erratic up to 1937, when 55,918 tons were made. This was the largest for any year since 1918.

Electric Percentage of Total Steel

Some rather interesting data have been compiled in Table II. This presents the percentage of electric steel of the total steel production of each of the 12 nations for 1937. In parentheses the corresponding percentages for each country in 1936 are added.

It will thus be seen that Italy, at 29.70 per cent, leads all countries in the proportion of electric of total steel made in 1937. It was 30 per cent in 1936. Sweden is second at 23.20 per cent in 1937 and 21.40 per cent in 1936. These are followed for 1937 by Canada, France, Japan, Germany and Poland. The data for the United States and the United Kingdom have changed but little.

It may be added that the total electric steel output of the 12 nations in 1937 at 3,250,499 tons was 2.84 per cent of the 114,530,000 tons as the total steel output of those countries.

This review plainly demonstrates that there has been, in the 25 years ending with 1937, a decided advance in this new industry. More or less experimental in its early days, it has advanced to an established phase of the steel industry. In the early years, Germany, Italy and France were the pioneers. American producers soon took up the cudgel, made decided progress, and today virtually lead the world in electric steel production.

The progress thus depicted is bound to continue as the demand for special alloy steels and alloys increases.

Table I—World Production of Electric Steel in Tons[1]

	1913	*1929*	*1934*	*1935*	*1936*	*1937*
United States	30,180	951,431	361,296	541,492	772,455	845,537
United Kingdom	22,000 (1915)	86,800	96,400	106,800	152,600	215,400
France	21,124	151,001	198,186	246,411	272,382	315,990
Saar	13,649	?	......	......	......	......
Germany	88,881	209,277[2]	236,207	362,465	498,537	700,369
Belgium	?	15,130	?	?	2,710	(3,200)[3]
Luxemburg	?	9,962	?	?	9,070	9,147
Japan	4,329 (1918)	52,797	149,270	173,042	(190,000)	(210,000)
Italy	36,948 (1917)	211,087	?	556,554	591,275	(610,000)
Poland	16,187	18,317	19,536	19,500	25,227	(30,000)
Sweden	2,276	112,702	164,599	186,268	206,221	253,810
Canada	449	52,761	23,160	33,499	34,953	55,918
India	?	?	?	?	38	1,128
Russia	No data					
	172,746	1,871,265	1,248,654	2,226,031	2,755,468	3,250,499

---

[1] Data are in both gross and metric tons. In most cases steel castings not included. [2] Crucible Steel, included from 1929 on, relatively small. [3] Data in parentheses are estimated.

Table II—Total Steel and Electric Steel Outputs Contrasted for 1937

	Total Steel	Electric Steel	Per Cent Electric
United States	50,570,000	845,537	1.67 (1.67)
United Kingdom	12,980,000	215,400	1.66 (1.29)
France	7,800,000	315,990	4.05 (4.12)
Germany (Saar)	19,540,000	700,369	3.58 (2.63)
Belgium	3,810,000	(3,200)	0.08 (0.08)
Luxemburg	2,470,000	9,147	0.37 (0.46)
Japan	5,720,000	(210,000)	3.67 (3.60)
Italy	2,050,000	(610,000)	29.70 (30.00)
Poland	1,430,000	(30,000)	2.10 (2.20)
Sweden	1,090,000	253,810	23.20 (21.40)
Canada	1,350,000	55,918	4.14 (3.23)
India	5,720,000	1,128	0.02

Observing Plastic Deformation With the Microgrid

by F. N. RHINES AND R. WARD

Ass't. Professor, Dept. of Metallurgy, and Research Ass't., Metals Research Laboratory, Carnegie Institute of Technology, Pittsburgh.

Technically, the nature of the process of plastic deformation in metals and alloys is a problem which interests research metallurgists and others. To study this problem, the observation of this plastic flow needs simple and effective laboratory methods. There have been various methods employed, each of which has certain limitations and advantages. The authors of this article describe the microgrid method which, despite some limitations, is claimed to possess the advantages of suitability and simplicity for certain studies—The Editors.

AS A CONSEQUENCE of the increasing interest in studies upon the nature of the process of plastic deformation in metals and alloys, the development of simpler and more effective laboratory methods for observing plastic flow appears desirable. Especially is there a need for procedures capable of revealing the microscale details of inter- and intra-crystalline movements. Various methods, typified by the works of Unckel,[1] Barrett,[2] and Seumel,[3] have been employed for the study of the course of plastic deformation within metal crystals. Each of the techniques used by these investigators has certain advantages and limitations.

The microgrid method, described here, may be used for studies of the same nature, and while it has definite limitations, some of which are not possessed by the older methods, it has the advantages of simplicity and of suitability for the detailed study of a relatively large region of a specimen.

Explanation of the Microgrid

The "microgrid" is simply a very fine grating ruled upon the polished and etched surface of a sample (Fig. 3) which is subsequently to be deformed. Such a grating may be ruled with the equipment commonly available in the metallography laboratory, namely, the mechanical stage of a microscope and a sharp pointed stylus. The grid is photographed before and after deformation, thus permitting the displacement of any point to be measured by a comparison of the photographs. Changes in dimensions are measured with a ruler, or more accurately with the vernier scale of the microscope stage, or with a microcomparator. Angular changes may be measured with a protractor, or with the divided scale of a rotating microscope stage.

In the present instance, the microgrids were ruled with the pointed diamond cutter of a microcharacter scratch hardness tester[4] mounted on the stage of a table microscope, Figs. 1 and 2. To assist in the compact and uniform spacing of the lines, a long lever arm was attached to the transverse motion gear of the mechanical stage. The polished and etched surface of the sample was moved under the cutter by means of the translation gear. At the end of each stroke the cutter was lifted and the sample moved back to its original position, displaced laterally a short distance by means of the accessory lever arm, and again drawn under the cutter. Two groups of parallel lines were ruled at right angles to give a cross grating.

The first requirement for the production of a good grid is a nearly perfect metallographic polish on the surface to be studied. Scratches, pits, and relief of the microconstituents may deflect the stylus and cause the lines of the grating to be irregular or to be partially obscured.

An Etch is Necessary

In order to reveal the location of the grains and microconstituents in the polished surface, an etch is usually required. Simple grain boundary etches are subject to the disadvantage that the pattern may be lost when the sample is deformed. For this reason the staining and contrast (etch pit type) etches are usually to be preferred. Too deep a pitting etch is likely to interfere with the ruling of the grating, however, and it is necessary to determine by trial the type

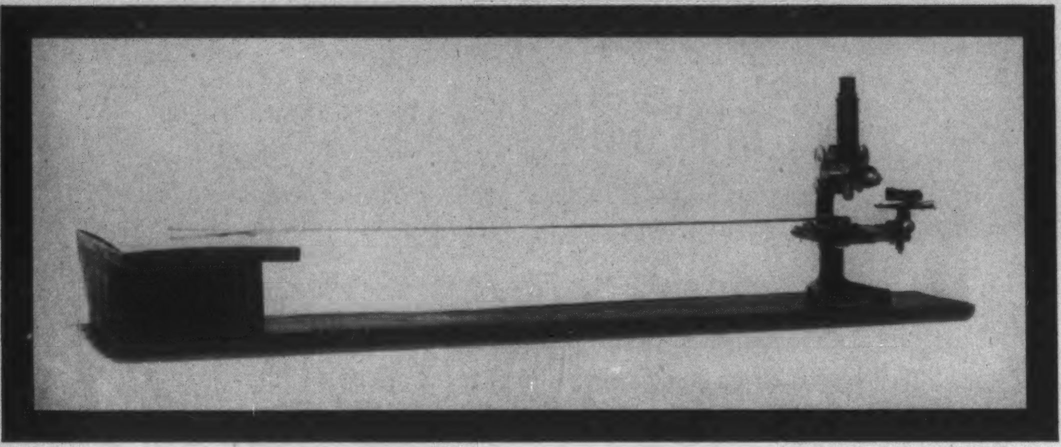


Fig. 1. Shows Microscope Equipped with Microcharacter and a Long Lever Arm on the Transverse Motion Gear of the Mechanical Stage.

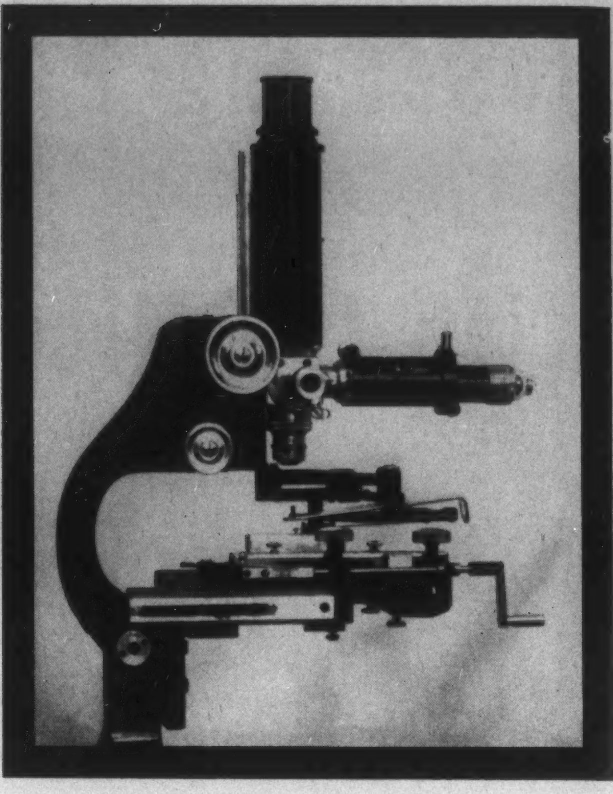


Fig. 2. Detail of the Microcharacter, Showing the Stylus in Contact with the Metal Specimen.

of etch best suited to each metal or alloy.

To produce the gratings shown in the accompanying micrographs, the samples were mounted with plasticene on a microscope slide which was placed in the mechanical stage of the microscope. An area for study was selected with the microscope and lines drawn across this region with the microcharacter stylus. Since it was desired to rule a large number of lines across each grain, their breadth was kept to the minimum consistent with the requirement that they should not be obliterated by the subsequent working. This was done by adjusting the load on the stylus point. For soft metals no load in addition to the weight of the stylus itself was used. The closest spacing of the lines practical under the conditions here described seems to be about 0.01 mm. It is important also that the lines be as straight as possible in order to prevent their overlapping. A rapid passage of the stylus over the surface of the sample was found to give the straightest lines.

In photographing the grid, it is advisable to use a lens of long focal length. The deformation of the sample inevitably gives rise to a rumpling of the surface, which in extreme cases prevents the preparation of a satisfactory photographic record. By the use of the greatest depth of focus of which the metallograph is capable at the desired magnification, the influence of the rumpling is minimized. Frequently, the pattern of the grid is more sharply delineated by the use of oblique or dark field illumination than by vertical illumination.

While the general nature of the plastic flow can be appraised by a visual comparison of pictures taken before and after deformation, an actual measurement of the dimensional changes is often very instructive. This is illustrated in one of the examples cited below. A microcomparator was used to measure the distance between adjacent intersections in the grid. Because photographic prints are subject to stretching, measurements were made directly upon the negatives.

How the Microgrid Method is Applied

The application of the microgrid method to the observation of the deformation of pure cast zinc is illustrated in Figs. 3 and 4. Here may be seen a complete single grain bordered by five or six others which are shown only in part. Several twin bands cross the grain, but these may be merely polishing twins. Fig. 3 shows the appearance before deformation, and Fig. 4 after a reduction in thickness of 10 per cent by rolling parallel to the plane of the picture. It is at once evident that the central grain has been rotated slightly with respect to its neighbors, that there is a more or less sharp bending of the lines of the grid at the grain boundary, and that the lines have in some localities spread and in others approached one another.

The local dimensional changes are better revealed by a plot of the changes in length of each line as measured between points of intersection on the grid (Fig. 5). The values plotted are the percentage change in linear dimensions versus the position in the grain. A separate curve has been drawn for each line of the grid. Positive values represent expan-

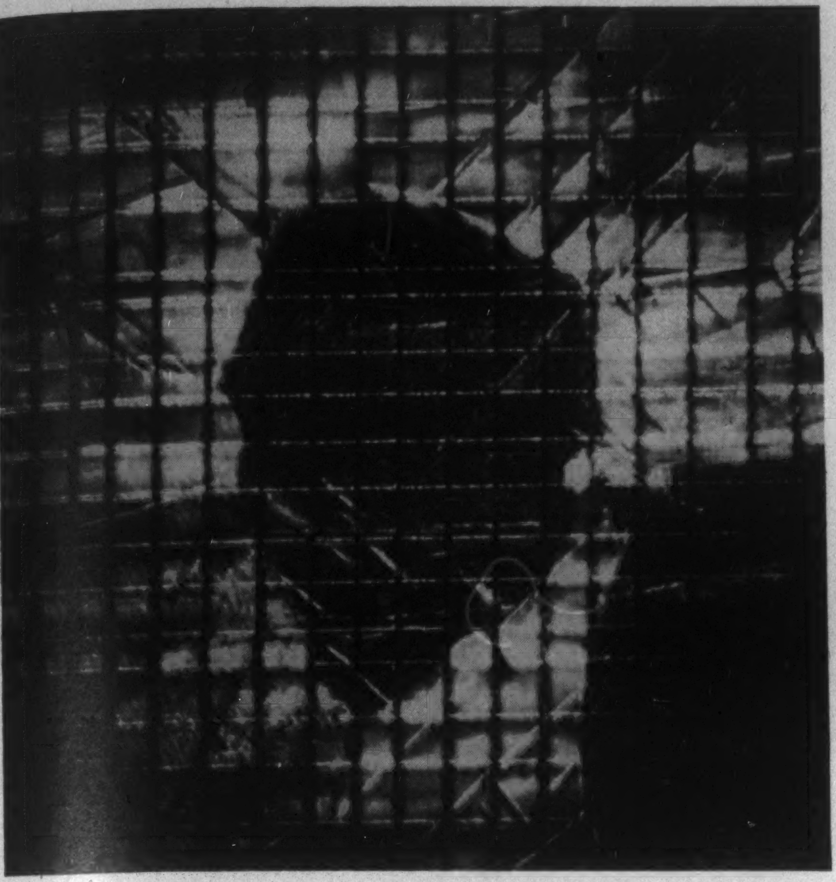


Fig. 3. Pure Cast Zinc Etched with Aqua Regia and Inscribed with a Microgrid. Magnification 85 X.

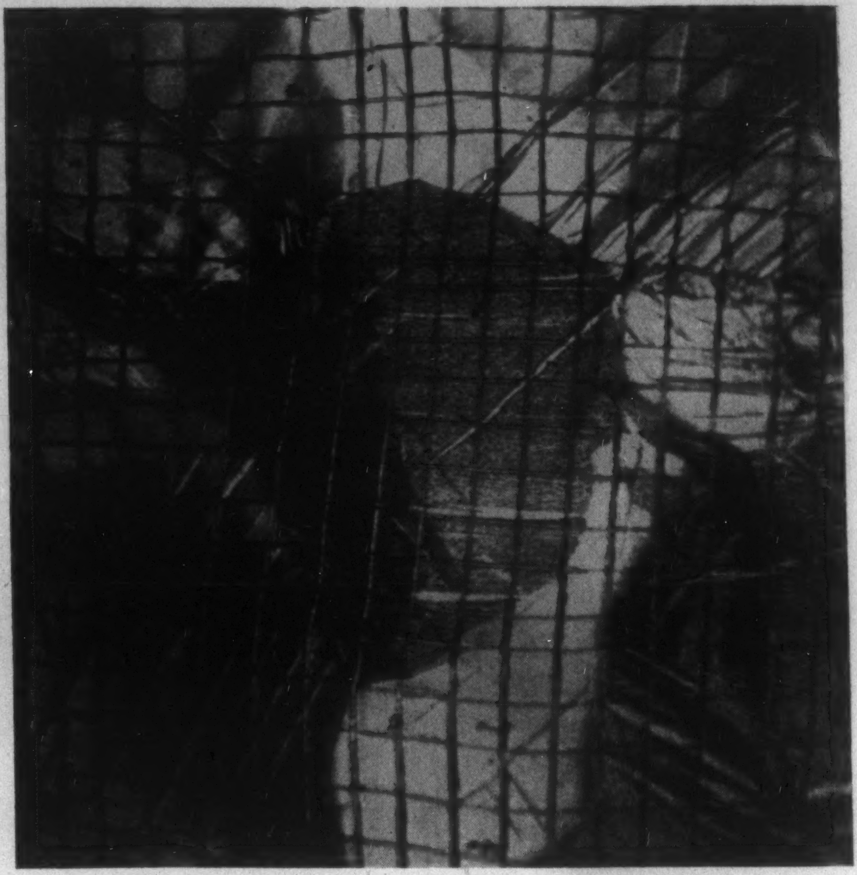


Fig. 4. Same as Fig. 3 Reduced 10 per cent in Thickness by Rolling in a Direction Parallel to the Plane of the Picture and Horizontal to the Page.

sions, negative values contractions. The lettered curves apply to lines perpendicular to the surface of rolling, while the numbered curves are derived from the lines lying in the direction of rolling.

It is obvious that the deformation is far from homogeneous. Indeed, almost every line crosses zones of both contraction and expansion, even those lying perpendicular to the surfaces of rolling. The influence of the neighboring grains in determining the nature of the local deformations is somewhat less evident, but a careful examination of the dimensional changes around the perimeter of the grain seems to reveal the presence of more or less abrupt alterations in the direction of the plastic flow in the regions immediately adjacent to two adjoining neighbors.

(A study of the orientation relationships of the undeformed grains of this sample might have been helpful in interpreting the results, but this line of investigation was made difficult by the smallness of the grain size.)

Presumably the grains lying wholly below the surface of polish are responsible for some of the irregularities observed. Their influence may be expected to be most apparent when the grain under observation is very shallow in depth. This behavior suggests that the nature of the grain boundary itself has less influence upon the plastic flow of the crystal than has the material of the adjoining grains. Finally, it should be noted that the twin bands have exerted no measurable influence upon the movements within the crystal. Perhaps this is because the bands are thin, or possibly because they do not extend far below the surface.

Problem of Polyphase Alloys

Polyphase alloys present a more complicated problem, since there is almost always a marked difference between the mechanical properties of adjacent phases. The microgrid is well adapted to the study of cases of this type, Figs. 6, 7, and 8. In these micrographs is shown the structure of a cast alloy of 88 per cent tin, 12 per cent antimony before deformation (Fig. 6), after a reduction of 5 per cent (Fig. 7), and after a reduction of 15 per cent (Fig. 8). As before, the sample was reduced by rolling on a surface parallel to the plane of the picture and in a direction horizontal to the page. The hard brittle particles of the β constituent (white) tend to crush, while the soft plastic tin solid solution flows in an irregular pattern about them. At the grain boundaries of the tin solid solution the sharp bending of the lines of the grid is apparent. In fact, some of the lines appear to be discontinuous at the grain boundaries, but a careful examination of the photographs proves this impression false. The inhomogeneous character of the deformation is again evident. It is interesting to note, however, that the movements revealed at a reduction of 5 per cent seem to have continued in the same general direction up to the 15 per cent reduction.

Principal Limitations

These two examples will serve to demonstrate some of the possible uses of the microgrid technique. They

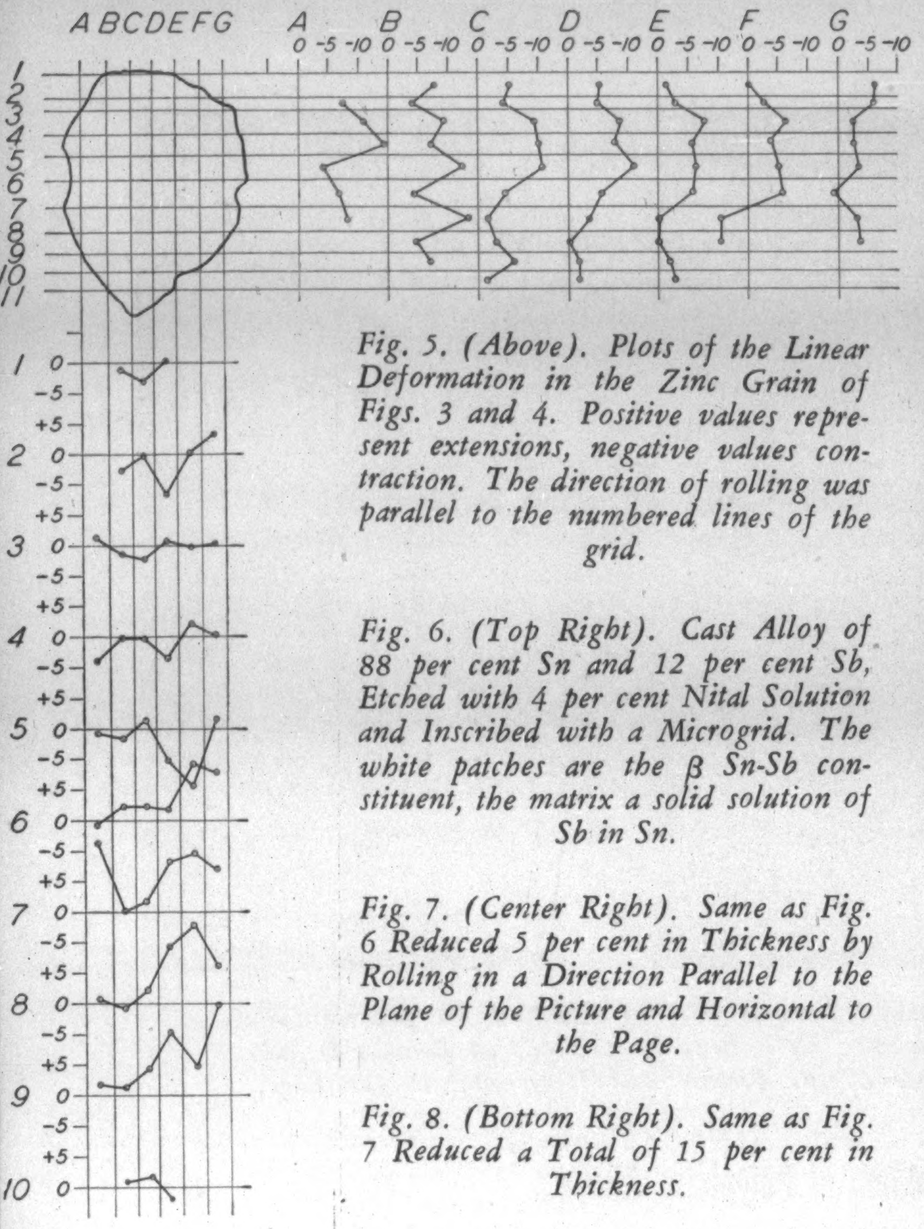


Fig. 5. (Above). Plots of the Linear Deformation in the Zinc Grain of Figs. 3 and 4. Positive values represent extensions, negative values contraction. The direction of rolling was parallel to the numbered lines of the grid.

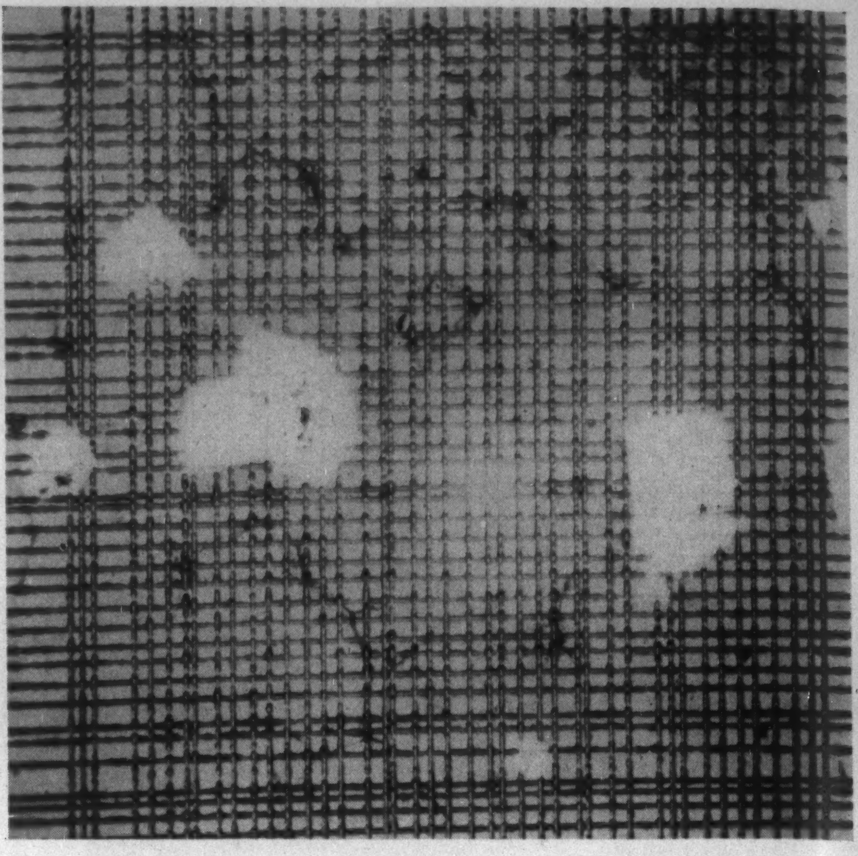


Fig. 6. (Top Right). Cast Alloy of 88 per cent Sn and 12 per cent Sb, Etched with 4 per cent Nital Solution and Inscribed with a Microgrid. The white patches are the β Sn-Sb constituent, the matrix a solid solution of Sb in Sn.

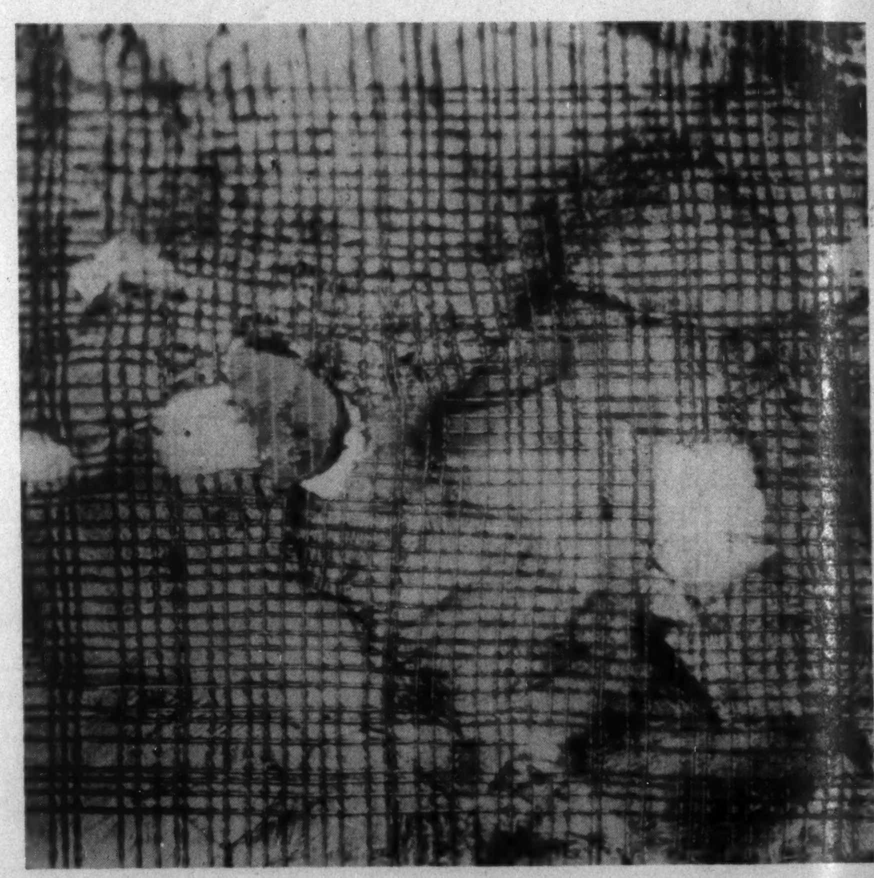


Fig. 7. (Center Right). Same as Fig. 6 Reduced 5 per cent in Thickness by Rolling in a Direction Parallel to the Plane of the Picture and Horizontal to the Page.

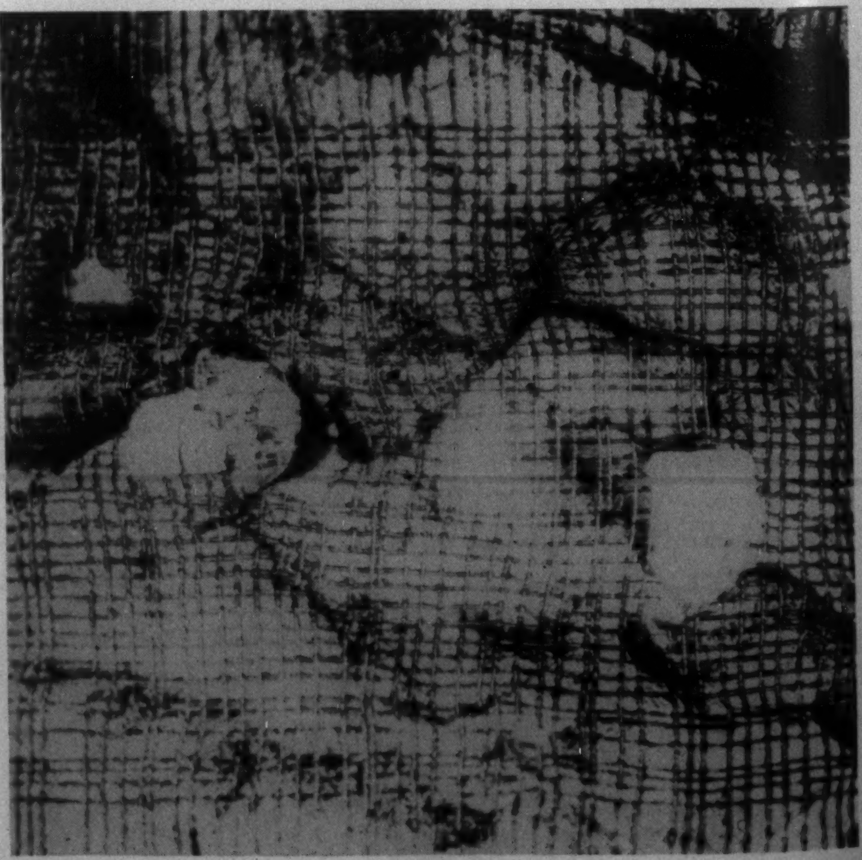


Fig. 8. (Bottom Right). Same as Fig. 7 Reduced a Total of 15 per cent in Thickness.

illustrate as well two of its principal limitations, namely, that the studies must be restricted to relatively small reductions made under conditions such that the grid will not be obliterated, and that the plastic behavior may be studied in a two dimensional external surface only. To some extent, the first of these limitations may be avoided by a direct visual examination of the samples. The microscope may be refocused to bring into view each section of the rumpled surface, but measurements of dimensional changes made in this way are subject to greater inaccuracies. A better alternative is to repolish the sample after the first few reductions and inscribe a new grid. A rough simulation of the conditions existing within the metal instead of at the surface might be attained by deforming the sample while it is embedded in a low melting matrix of similar mechanical characteristics or by rolling the grid surface in contact with a polished surface of another piece of the same material. If desired, the plastic flow in a direction perpendicular to the surface of polish may be measured by the use of a microscope equipped with a calibrated vernier scale on the focusing gear. This will allow the plotting of a three dimensional description of the flow in the surface.

References

[1] Hermann Unckel. *J. Inst. Metals,* Vol. 61, 1937, page 171.
[2] C. S. Barrett. *Metals Tech.,* Vol. 5, Oct. 1938, T. P. 977.
[3] Gerhard Seumel. *Z. Krist.,* Vol. 93, 1936, p. 249.
[4] C. H. Bierbaum. *Trans.* A.I.M.E., Vol. 69, 1923, p. 972.

COMMENTS ON CARBON-LINED BLAST FURNACES

By F. J. VOSBURGH

Consolidated Carbon Division,
National Carbon Co., Inc., New York

The article by Dr. Singer is a most interesting one and blast furnace operators should find in it much food for thought. If anything, Dr. Singer has understated rather than overstated the facts. The writer has just returned from a six weeks' visit in Germany made with one object in mind, to see a carbon lining installed in a blast furnace.

It was claimed by men entirely familiar with the situation that 75 per cent of all blast furnaces in Germany are carbon lined and that even a higher percentage of the iron is made in carbon lined furnaces. It was claimed that one furnace has produced over 3,000,000 tons from a carbon lining and that several other furnaces will undoubtedly go over that figure. The writer saw two operating furnaces, both of which were blown in in 1928, one of which has produced over 2,000,000 tons, and the other, which has been the standby furnace during that period and therefore has not been operated as consistently, has produced over 1,500,000 tons. No one in the operating department has any doubt but that both furnaces will produce iron for another five years.

Dr. Singer mentions carbon lining as being used in Sweden and Russia as well as Germany. To that list should be added Belgium and Italy.

Dr. Singer has mentioned one fact, but has not emphasized it particularly. The writer was told that in using carbon linings, no cooling plates are used in any part of the furnace that is carbon lined and the furnace he saw lined had no plates up to the mantle. All cooling was to be on the outside of the shell.

Three types of carbon linings are generally used:

First—Carbon brick in the standard blast furnace brick sizes: I was told that the carbon brick lining is more satisfactory than the ceramic lining, since the carbon brick are unattacked by either the slag or metal, but such a lining still has the other weaknesses of the small size ceramic brick lining. This type of lining is by far the most prevalent. Most of the ceramic manufacturers make or have made carbon brick and many of the blast furnace operators made their own brick about as described by Dr. Singer.

Second—The tamped, or monolithic, carbon lining: It is claimed that this type of lining has been in use since 1929, but the writer believes reference is made to the use of carbon paste as a patching material for a carbon brick lining of the type described above. It was at a much later date that a complete monolithic lining was actually tried, and there appears to be some question of the actual results obtained. Regardless of that, there are a considerable number of furnaces so lined and only the final results will tell the story. The writer, as a carbon manufacturer, questions whether masses of carbon as large as are required in blast furnace construction can be tamped into place and baked out in the uncontrollable temperature of a blast furnace without developing cracks that will ultimately cause trouble.

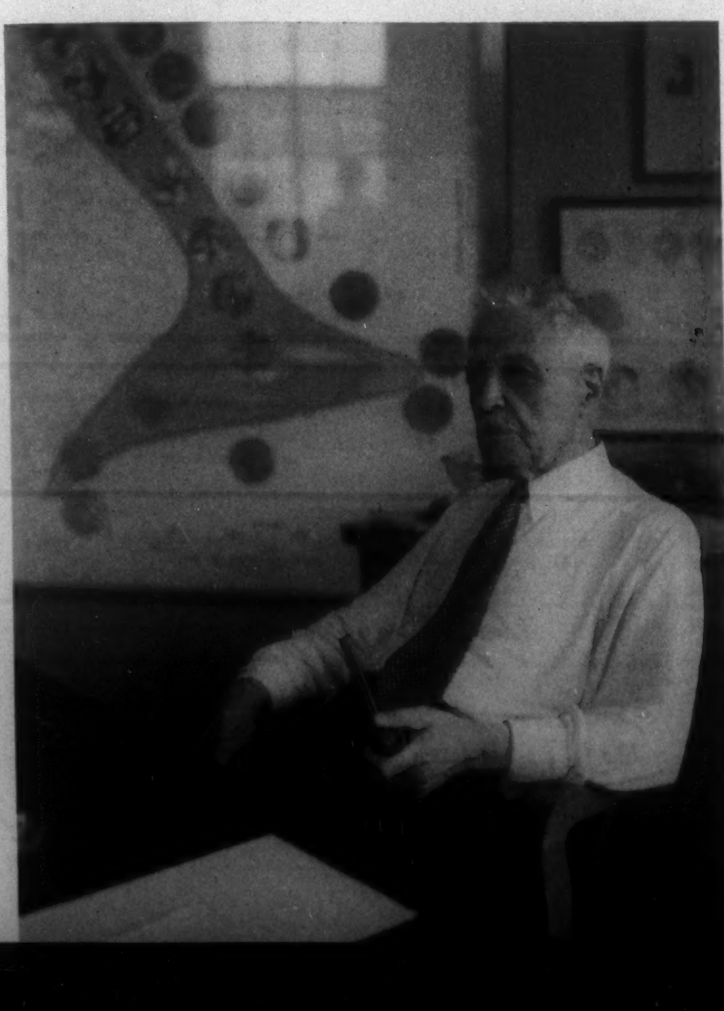
Third—The largest carbon manufacturer in Germany has developed a furnace lining made up of carbon blocks which in general approximate 2 x 2 x 2 ft. and weigh about 800 lbs. each, some being considerably larger up to 24 x 30 x 80 to 90 in. down to 15 x 18 x 18 in. These blocks are baked in furnaces such as are generally used for baking carbon electrodes and at a minimum temperature of 1100 deg. C. The blocks are then cut and ground to such exact dimensions that, when they are laid up in the furnaces, the joints are absolutely tight. It was this type of lining that the writer saw installed in an 800-ton furnace and too much emphasis cannot be placed on the splendid machine work which had been done and which permitted the installation in the furnace with absolutely no difficulties of any sort, each piece matching up exactly with the next one.

It is rather remarkable that in Germany carbon has been used as blast furnace lining, as the Germans say "from the beginning," while to the writer's knowledge carbon has never been so used in the United States or Canada.

The writer, because of his recent contact with the situation in Germany and because of his great interest in the subject, appreciates particularly the opportunity given him to comment on Dr. Singer's article.

●

A photograph of the late Doctor Sauveur made in the summer of 1938 at the Research Laboratory of the United States Steel Corp., Kearny, N. J. (Courtesy: Dr. John Johnston).



Editorial

(Continued from page 103)

poses, are even accepted as news—that is, the lay reader is so interested in general advances, that even these pseudo-advances get attention. Too little attention is paid to publicizing the advances that can be proven by statistics, and which are no longer news to the technician. These are still news to the general public.

The historian's angle would apply to many cases. As we write this the daily papers note, as important news, the fact that a relatively small rail order was placed, after a long dearth of such orders. Why did the "public relations" departments of the steel companies and the railroads not have already prepared and filed in the "morgue," an account of the development of slow-cooled rails, detector cars, end hardening, welding of long lengths and the other advances that make today's rails safer and more enduring. A "release" of that type, tied to the spot news of a rail order, would have so much appeal to the general public that it would doubtless be printed in the majority of newspapers, and it would help give the public a better appreciation of what "big metallurgical business" and the railroads have been doing for the safety of the public. Provable facts of this nature, when they are assimilated by the public, go a long way toward neutralizing the effect of administration name-calling.—H. W. G.

The Volatile Industry

Poor's Survey of the Steel and Iron industry of Feb. 15 points out that prospects are good enough to justify recommendations to hold or to buy stocks of the industry, in spite of the deficits chalked up against so many of the firms for 1938. One phrase of this survey caught our eye, the "outlook of this volatile industry."

We may not like the term, but we have to admit that it's an apt one to appraise the difficulties under which the metallurgist has to work, in contrast to those in the candy, cosmetic and cigarette industries. There may be virtue in the "C," for the making of steel for cans is the least volatile phase of the steel mill's work.—H.W.G.

The Essentials

In a description of the U. S. Naval Research Laboratory, Dr. Gunn remarks, "The important thing in scientific research is to have an instinct for the essential."

That this might be said with truth about many other lines of activity does not detract from its special applicability to research. Even if one is not born with this instinct well developed, it can be developed to a helpful degree by stopping more often to think of one's problems in terms of the essentials.

Research directors and research workers might well frame this motto and hang it over their desks and work tables.—H.W.G.

A Few Chuckles

Pure Metals

To the Editor: Here is how metals are classified on the air. "Professor Quiz," in his radio program of Saturday, Feb. 26th, asked a contestant:

> "Which of the following are *pure* metals: Copper, brass, bronze, cast iron and aluminum?"
>
> Answer by contestant: "Copper and aluminum."
>
> States the professor with gusto and deadly certainty: "Wrong. Copper, aluminum and cast iron. Cast iron is pure iron."

H. S. Blumberg.

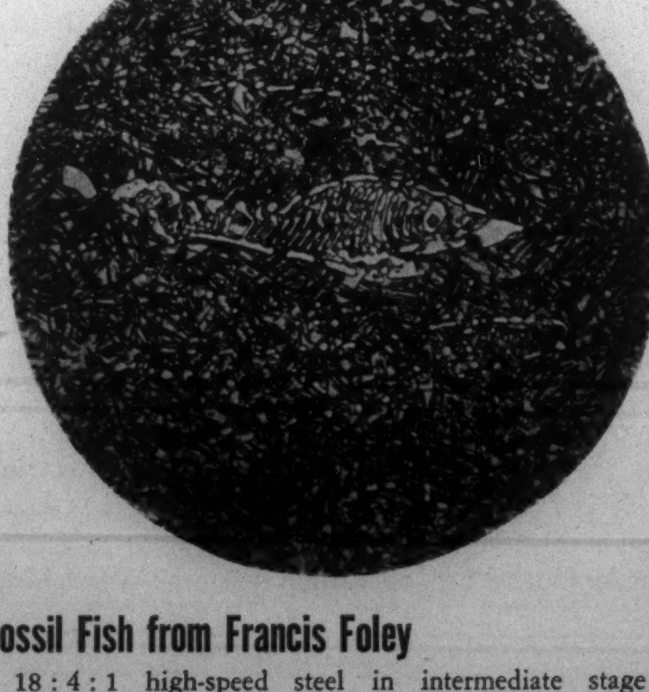
428 Lafayette St., New York.

N.B.: Because of the contestant's "wrong answer," he lost first prize.

Non-Metallurgical "Creep"

The "chuckle" on "Non-Metallurgical Creep" in our March issue, page 102, is pointless because of the omission of the caption to the illustration. This should have been: *"Our Bureau of Standards gives it a creep-stress rating of 1.036!"*

We are glad to pass this on in the hope that our readers will not object too strongly to receiving these jokes in installments.



Fossil Fish from Francis Foley

18 : 4 : 1 high-speed steel in intermediate stage of working.—H.W.G.